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The articles appearing on pages 1 to 228 inclusive of this issue of TRANSACTIONS comprise nine of the seventeen papers which constituted the Symposium on the Plastic Working of Metals presented at the Eighteenth Annual Convention of the Society held in Cleveland, October 19 to 23, 1936. The first eight papers of the symposium were published in the December 1936 issue of TRANSACTIONS.

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FACTORS RELATING TO THE PRODUCTION OF DROP AND HAMMER FORGINGS

By Adam M. Steever

Abstract

This paper discusses the various types of equipment used for the production of drop and hammer forgings. It touches on the importance of die manufacture, maximum forging temperatures, finishing temperatures and grain flow.

Tables are given showing maximum forging temperatures for some types of metals.

Curves are presented showing steam consumption of steam hammers under varying conditions.

Metallurgical factors are discussed that must be considered and controlled for the fabrication of satisfactory forgings.

THE development of processes to transform a bar or billet of steel by forging into an essential part of a locomotive, marine or gas engine, an automobile, or any of the other countless mechanisms, is an achievement that has helped our civilization forward. It is not a laboratory development but a gradual evolution at the forge and thereby still retains the dignity of an art.

The manufacturing methods and processes used in the production of forgings probably embrace a wider and more diversified field of activity than any other manufacture of metallic shapes, because it covers all types and kinds of forgings ranging in weight from a fraction of an ounce to hundreds of pounds per forging.

Hammering[1] was the first method employed by man in shaping

[1]The Making, Shaping and Treating of Steel, Camp and Francis, Carnegie Steel Co.

A paper presented as part of the Symposium on the Plastic Working of Metals, Eighteenth Annual Convention of the Society held in Cleveland, October 19 to 23, 1936. The author, Adam M. Steever, is superintendent, Columbia Tool Steel Co., Chicago Heights, Ill. Manuscript received June 25, 1936.

the metals. The first forging was done by hand hammers wielded by the workmen. The first power hammer, known by the name of tilt hammer, was built in England and was a crude affair compared with the steam hammers now used. Fig. 1. It consisted of a beam of wood hinged at one end and provided with an iron hammer head at the other. At an intermediate point, engaging cams on a revolving shaft alternately raised the free end and allowed it to fall on a

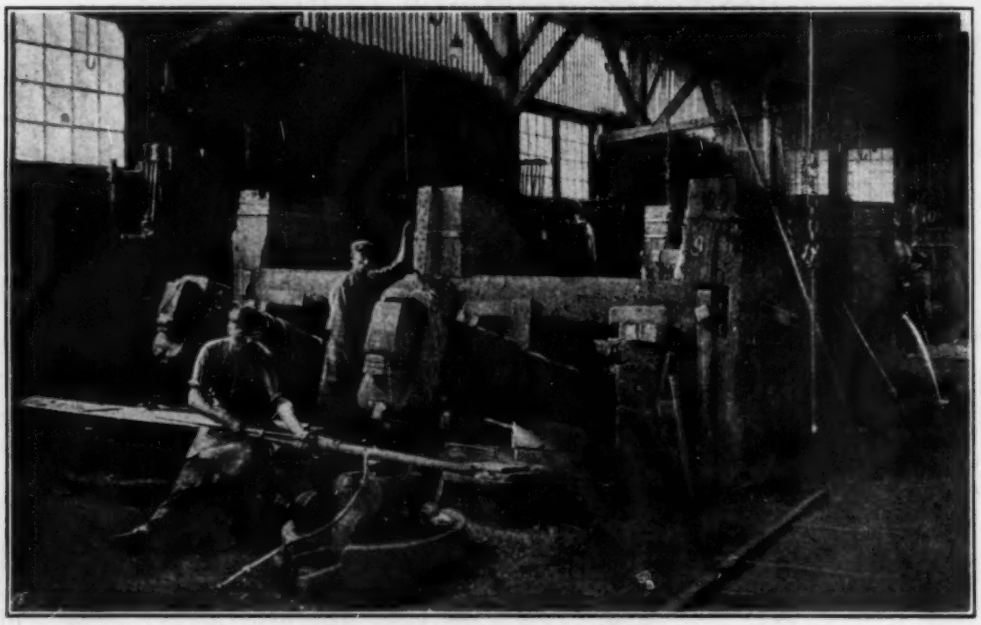

Fig. 1—Water Power Hammer.

bottom die fixed upon a suitable foundation. Thus the top die could be parallel to the bottom die in only one position and the larger the piece to be forged, the less power there was available to forge it.

The first steam hammer was built in France in 1842. It consisted of a two-piece frame constructed so as to support directly over a die or anvil a steam cylinder, to the piston rod of which was attached a tup or hammer head. Fig. 2. By admitting steam into the cylinder below the piston, the hammer was raised for a distance equal to the stroke of the cylinder and then allowed to drop upon the anvil or bottom die. This hammer had the advantage of always keeping the top and bottom dies parallel, but was still lacking in one important particular. Its power being derived from the inertia of the falling tup or ram, the hammer had the least power when it was most needed, that is, when pieces of large diameter or of great thickness were being worked. This fault in the single acting hammer was

corrected by the invention of the double acting hammer in which steam is admitted at the top of the piston and employed on the downward stroke as well as for lifting the ram. The first double acting hammer was built at Midvale, Pennsylvania, in 1888.

The principles of the hammer are that of an instantaneous application of pressure applied to a relatively small area. The strains set up in the metal are compressive and take place in a vertical direc-

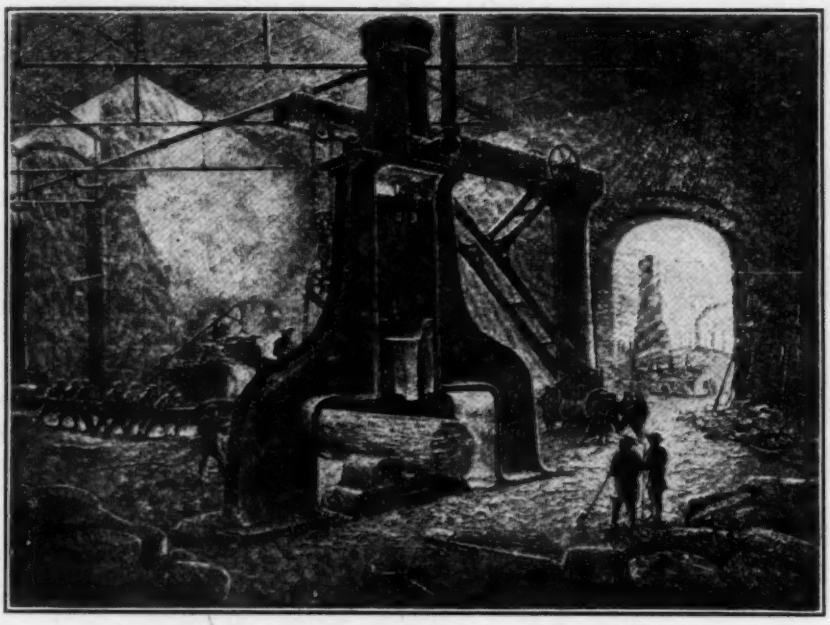

Fig. 2—Artist's Conception of The First Steam Hammer.

tion in a region below the area subjected to the force of the blow. The crowding of the metal into one region, however, causes a small portion of the blow to be transmitted in horizontal directions. The suddenness of the blow tends to localize the effect and confine the refinement to the exterior. This fact results in a high degree of refinement, provided the amount of reduction is great or the section worked is a thin one, and is one of the reasons why it is possible to make some hammered material superior to rolled material. The resistance of the metal to deformation under shock, combined with the intermittent action of the hammer, makes shaping by forging a slow process compared with rolling.

Forgings are produced by impact. The energy delivered by the ram to the forging in the form of useful work is dependent on falling mass, the velocity of the falling mass and the period of time in which the downward motion is stopped. By control of these factors

various forging capacities are regulated. There are two types of hammers used in forging: The steam or double acting hammer, and the board or gravity hammer.

In a gravity hammer, such as a board drop, the available energy is equivalent to the total falling weight times the height of the fall, less friction. The elimination of friction is the one means of increasing the effective blow.

In a double acting hammer, such as a steam drop, the available energy is equivalent to the total falling weight times the height of that free fall which will give the corresponding velocity at impact, minus the friction. Since the available energy increases with the square of the velocity, the limiting factor becomes the velocity of impact, which will cause metal to flow rather than be destroyed. To illustrate: a 5000-pound steam drop, with an impact velocity of 34.4 feet per second, has approximately 91,800 foot-pounds available energy. The same energy in a 50-pound ram would require an impact velocity of approximately 344 feet per second. Such a reduction in the time element for reaction on a forging would make the deformation one of destruction rather than one of plastic flow.

Steam drop hammers are rated by the nominal weight of the ram, piston rod and piston. This constitutes the falling weight. The actual scale weight of these parts usually exceeds the rated size. Steam hammers range in size from 400 to 25,000 pounds.

Table I gives the standard dimensions as furnished by a well known company on their line of steam hammers. These data correspond closely to the standard dimensions of the other makes of steam hammers, except in falling weight, which is greater on this particular make.

The anvil block is cast in a ratio of 20 to 1 to the falling weight, except in case of large hammers. One piece anvils can be furnished for hammers up to 12,000 pounds. The function of the anvil is to convert the available energy of a hammer into useful work. When the anvil block ratio falls under 20 per cent, the efficiency begins to fall off, all other variables being equal.

The sow block is dovetailed into the anvil on one side, keyed on the other and acts as a renewal holder for the bottom die. It should be made of heat treated alloy steel so that the dies will not seat into it. Into this block the stationary die is keyed. Two side frames are carried on the anvil and tied together both top and bottom, front and back, to prevent spreading. The overhang of the

frame, front and back, permits a wedge adjustment for the front and back. Also a lateral movement of the frame is made by wedges. This type constitutes the full wedge adjustment. These wedges assist in die setting and take up wear in the frame and anvil as it occurs. Other types have the side frames pocketed in the anvil to prevent lateral movement.

A steel tie plate spaces the frames at the top and holds the cylinder in alignment with them by means of an interlocking construction. The cylinder is made with the valve chest, bottom flanged and the stuffing boxes for the steam and exhaust pipes, all cast integral and held together and reinforced by means of heavy ribbing.

Two ram guides are set in side frames to take up wear and serve as a guide for the ram. Adjustment is made by means of a forged steel wedge at the back of the guide pocket. The guides are held in place by guide bolts running through the side frame.

The ram into which the movable die is keyed should be of forged steel and treated to such degree of hardness that the dies will not seat into it. The ram is moved by an alloy steel, heat treated rod attached to it, carrying a head on upper end which operates in the steam cylinder. The piston rings are from steel of special analysis.

The admission or throttle valve is of the rotary or Corlis type, the motion valve or distributing valve of the piston type, moving vertically. The motion valve gearing is of the combination control type, the position of the valve being regulated by the movement of a foot treadle and also automatically by the position of the ram.

A spring buffer type safety device is mounted on the cylinder by means of through bolts connected to lugs cast on cylinder body. This device protects the cylinder head from overtravel of the piston.

For lubrication, oil is introduced into the cylinder ahead of the throttle valve by means of either an individual lubricator or a central system.

The illustration, Fig. 3, shows a steam drop hammer. Fig. 4 shows a double frame forging hammer. Fig. 5 shows a modern board drop forging hammer. Fig. 6 shows an anvil mounted on a typical foundation design.

Steam supplied to the hammers should be dry and there is no objection to a small amount of superheat. The amount of steam required for hammer operation depends, to a great extent, on local operating conditions. The various factors affecting amount of steam required are length and force of blow, number of blows per

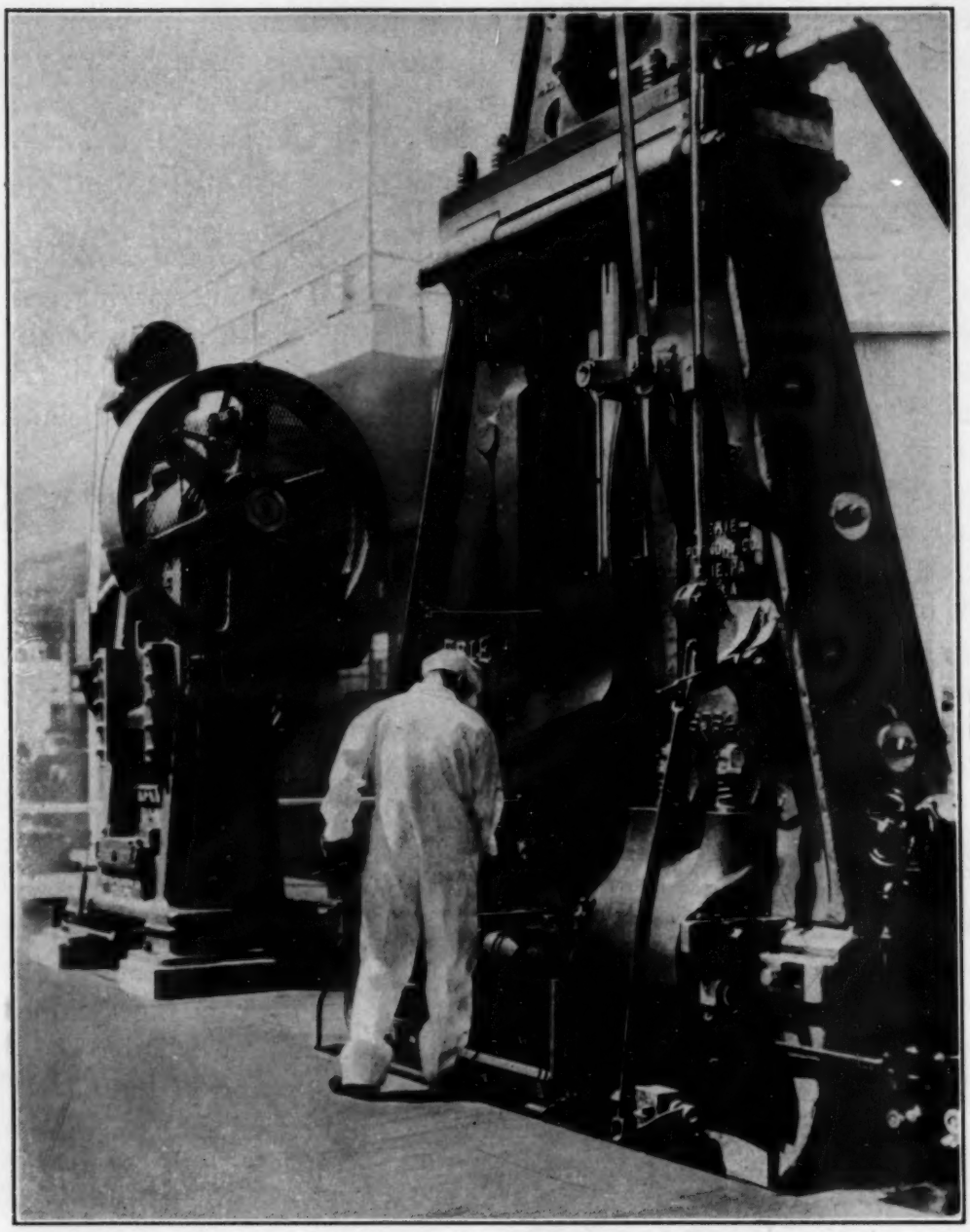

Fig. 3—Modern Steam Drop Forging Hammer.

minute, number of minutes of actual operation and the mechanical condition of the hammer.

The curve, Fig. 7, here given, showing the steam consumption of hammers of various sizes, was plotted from data obtained in a large number of tests under widely varying conditions. These curves show the great waste of steam which occurs when hammers are operated with valves, cylinder bore and piston rings badly worn.

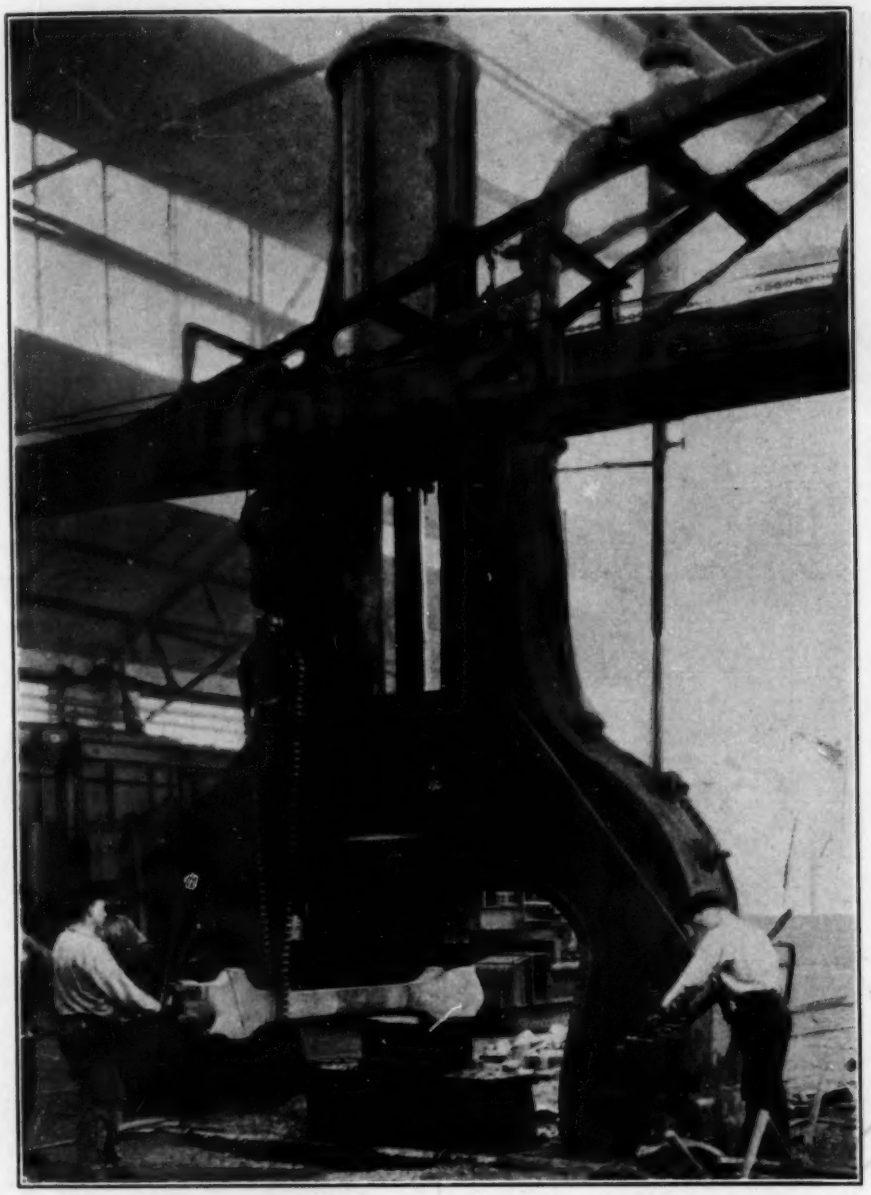

Fig. 4—Modern Steam Double Frame Forging Hammer.

The space in a hammer depends upon the length of the piston rod and depth to which it extends in the ram. However, the minimum die space is usually between 12 and 16 inches.

The open frame steam hammer is similar to the steam drop with few exceptions. Namely, the frame is cast as a single unit in form of a "C." The cylinder is mounted on top of the frame. The anvil is not directly connected to the frame, it being mounted as a separate unit on the foundation. Operating principles and mechanism are

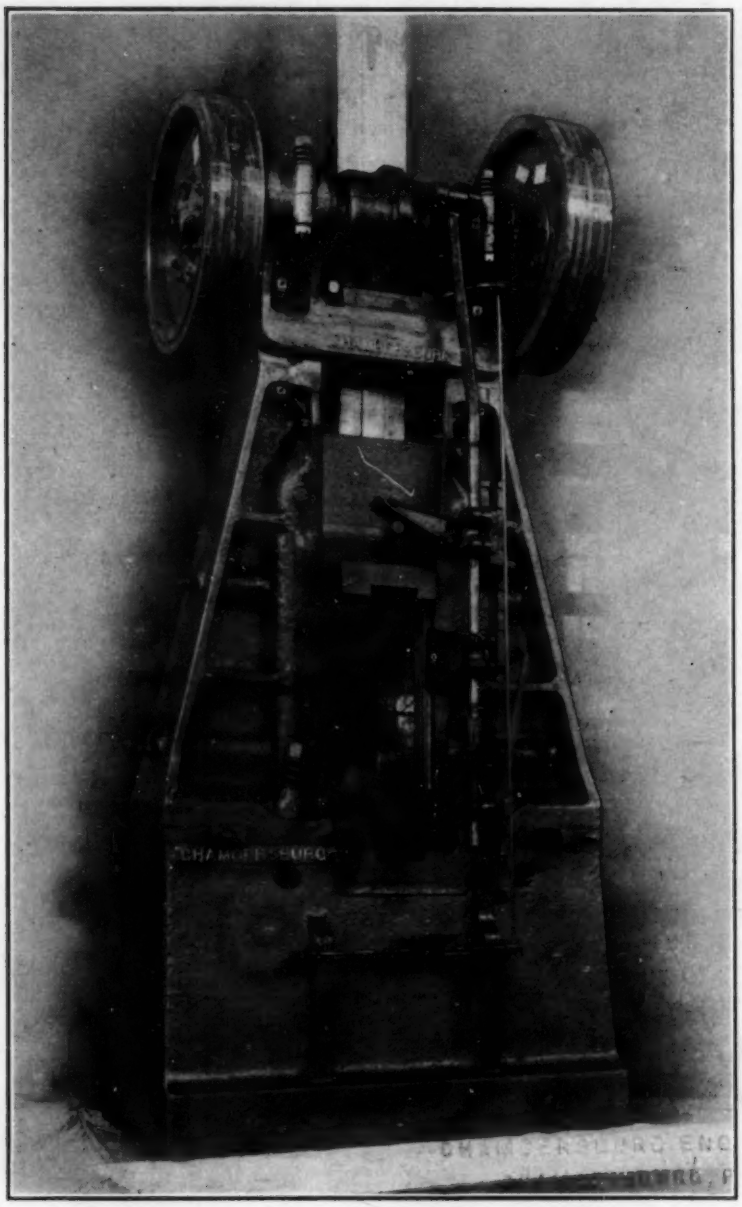

Fig. 5—Modern Board Drop Forging Hammer.

similar to the steam drop hammer with the exception that the open frame hammer is usually operated by means of a hand lever.

Air Drop Hammer

Steam drop hammers can be operated just as well on compressed air. From the standpoint of operation, either medium is satisfactory, the choice resolving itself into a question of economy. When compressed air is used, it should be free from moisture. Headers must

Table 1

Specifications of Steam Hammers

Size Pounds	Diameter of Cylinder Inches	Stroke Inches	Face of Ram Between Guides Inches	Face of Ram Front to Back Inches	Length of Ram in Guides Inches	Actual Weight Ram, Rod and Piston	Weight of Upper Works	Total Weight
400	5½	30	12	8	14	450	6000	14000
600	6	30	14	12	16	670	9000	21000
800	7	30	15	13	18	1000	10000	26000
1000	8	36	16	14	23	1400	15000	35000
1500	10	42	20	16	27	2200	20000	50000
2000	11	45	23	17	28	2800	24000	64000
2500	12	50	24	24	28	3400	30000	80000
3000	14	50	26	26	35	4600	32000	92000
4000	15	50	28	30	36	5200	39000	119000
5000	17	50	28	30	36	6600	51000	151000
6000	17	50	28	34	36	7200	52000	172000
8000	19	50	30	38	40	8525	64000	224000
10000	21	50	30	38	40	10650	66000	266000
12000	23	50	30	40	44	12400	70000	310000

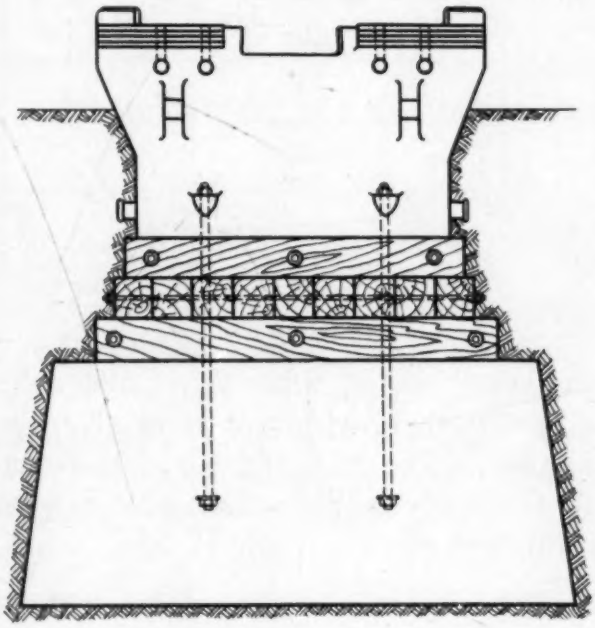

Fig. 6—Typical Hammer Foundation.

be installed between the air compressors and the hammers, of sufficient capacity to supply air to the hammers at a constant pressure.

Air compressors are usually rated in volume per minute, and hammers operated on compressed air are therefore on the same basis. Tests show that hammers in average operating condition require from 18 to 25 cubic feet of free air per minute for each 100 pounds of rated size.

Board Drop Hammers

The use of board drop hammers has increased greatly in recent years. Some of the advantages of the board over the steam hammer are, no expensive boiler plant with its high maintenance costs is required, operators do not require such high degree of skill and can therefore be trained more quickly, and also in general lower maintenance costs.

Recent design of these hammers has given a more rugged hammer with a size range up to 7500 pounds. Board hammers are oper-

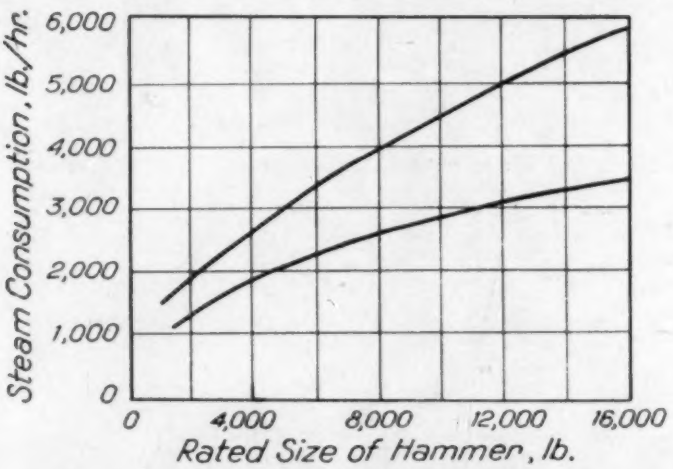

Fig. 7—Curve of Steam Consumption. Upper Curve—Hammer in Poor Condition. Lower Curve—Hammer in Good Condition.

ated direct from electric power, either direct motor driven as on the newer types or by line shaft with pulleys to give the proper speed of the eccentric rolls. The standard board hammer specifications from 1000 to 7500 pounds falling weight of a representative board hammer manufacturer is given in Table II.

Table II
Board Hammer Specifications

Falling Weight	Between Guides	Extreme Fall Without Dies	Ram Front to Back	Cap Front to Back	Floor Space	Pulleys	R.P.M. Pulleys
1000	17	48	14	21	35x 72	36x 6	130
1500	18	51	16	23	39x 80½	42x 7	120
2000	19	55	20	26	42x 88	48x 8	110
2500	20½	57	21	27	45x 92½	48x 9	95
3000	23	59	23	30	46x 99	54x11	85
3500	23	59	24	30	46x 99	54x11	83
4000	25	61	25	32	54x112	62x14	80
5000	27¼	63	27	37	63x124	72x17	72
6000	29	66	29	40	72x128	72x19	60
7500	30½	70	32	42	81x132	78x21	55

The foundation, anvil and sow block of a board hammer are essentially the same as the steam drop hammer. The anvil is cast in a ratio of 20 to 1 of the falling weight. The top of anvil block has stepped and tongued frame seats on each side. A breather hole under each seat prevents accumulation of scale that would interfere with the fit. Wearing plates on the sides of the tongue can be renewed without disturbing any other part of the hammer. Lateral die adjustment is made by forged steel adjusting wedges which shift the entire upper works. The frame is held to the anvil by four large anvil bolts.

The lifting mechanism is entirely different in a board hammer. One or more boards are keyed into the top of the hammer ram, and two rollers at the top of the hammer roll the boards upwards. When the head has reached the length desired, a trip on the side of the hammer head operates a latch rod, which in turn spreads the rolls apart and allows the hammer head, movable die and board to fall freely on the work below. As the hammer reaches the bottom of its stroke, the latch rod throws the rolls together again and they roll the board to the top of the stroke. The operation is controlled by a foot lever.

The guides in the side frames are adjusted similar to the steam hammer guides. A very rugged interlocking tie plate maintains frame alignment in all planes and protects the head assembly from twisting, strains and accidental overtravel of the ram. The rolls are mounted on large heat treated alloy steel shafts, kept in parallel alignment by roller bearings. Hammer boards are made from clear, straight grained maple, free from all knots. Many attempts have been made to substitute paper, fibre and other materials but nothing as yet has been found which will satisfactorily withstand the severe crushing pressure of the rolls.

Die Blocks

Die blocks for either drop or hammer forgings must withstand heavy stresses under impact.[2] They must be made of a composition that will minimize abrasion, warping and fire checking. The majority of die blocks are made either by the electric furnace or open-hearth process. Their composition and heat treatment is given in

[2]"Practice for the Heat Treatment of Plain Carbon and Alloy Steel Die Blocks," *Metals Handbook*, American Society for Metals.

the METALS HANDBOOK, American Society for Metals, under the subject of Die Blocks and will not be discussed here.

Die blocks used for hammer forgings are usually in one plane, that is, flat surfaces. Blocks used for drop forging are of various designs due to the shape of the forging being made.

There are many factors that govern the life of a die impression: Temperature of metal being forged; size of hammer; type of forging steel; manufacturing tolerance; mechanical damage and method of holding dies in hammer as to proper bearing, etc.

The making of dies and tools is an important factor in the production of drop forgings. This process is commonly known as

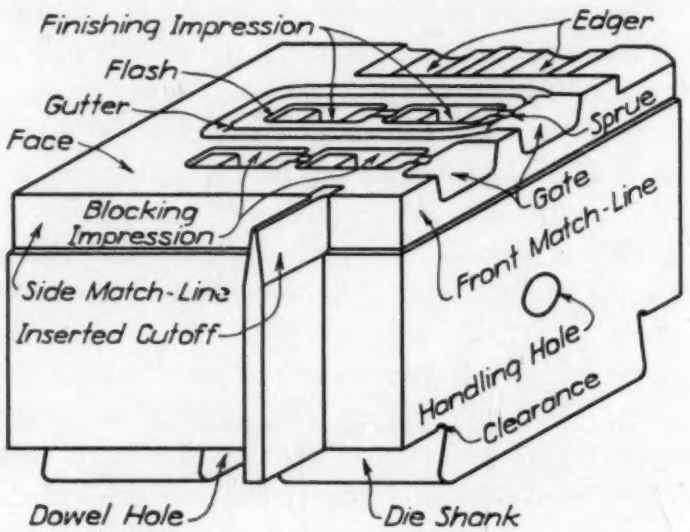

Fig. 8—Diagram of Drop Forging Die.

die sinking. The purpose of drop forge dies, when placed in air, steam or board hammers as well as upsetters and presses, is to produce forgings in quantities which are, practically speaking, duplicates of each other.

To properly construct dies it is necessary to have definite information regarding the piece to be produced, viz., shape of the piece to be forged, size of steel to be used, grade of steel, size of hammer, quantity of forgings and future prospects. The shape of the piece and intricacy of design determines the size of the die blocks. After the die blocks have been selected, they are drilled on each end, and these holes are used for clamping on the planer, when machining, and to aid in handling the succeeding operations.

Blocks are usually made of an alloy steel, carefully heat treated, the hardness varying to suit the particular requirements of the part

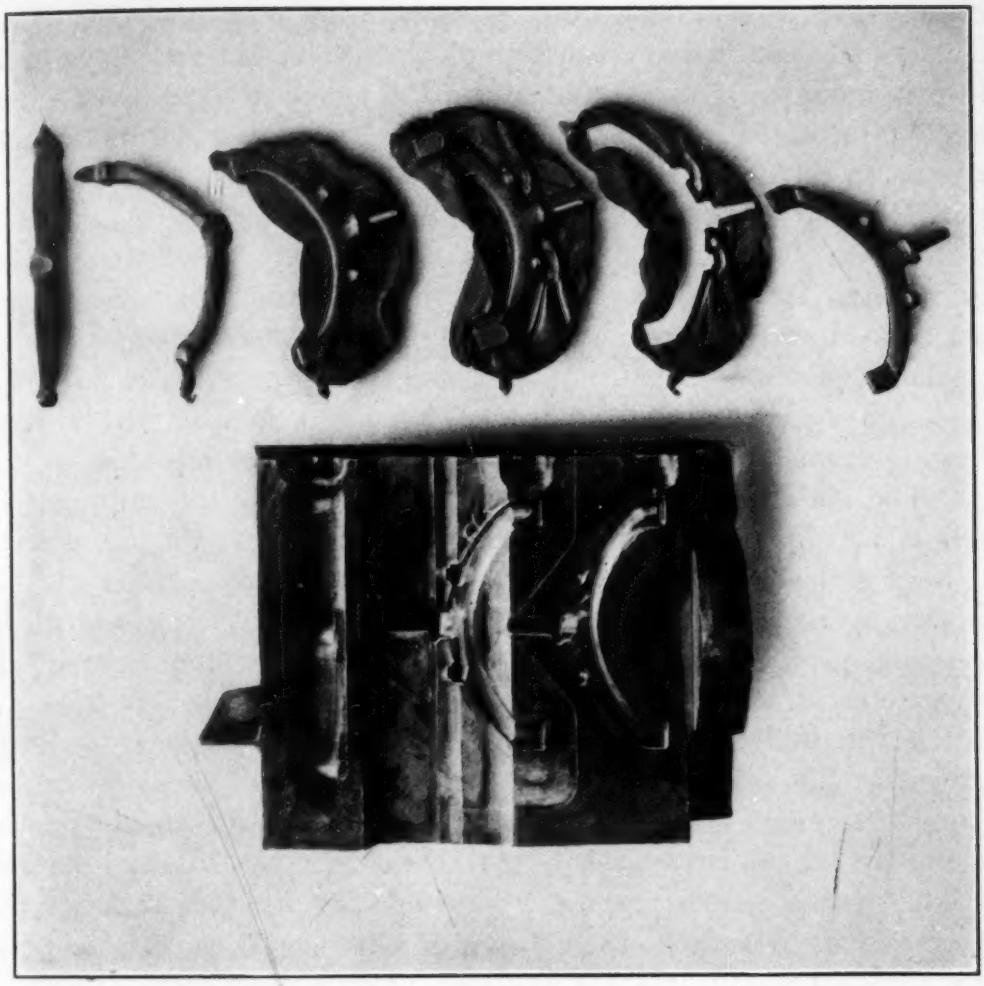

Fig. 9—Photo of Drop Forging Die Showing Various Operations and Impressions.

to be made. After drilling, the blocks are planed, shanked and faced with a matched edge on each block laid out so that all measurements on each block can be taken from these two points, thus bringing the impressions in perfect alignment in the forging operation.

The parting line of the dies, which is the fitting of the top and bottom die together, may be either flat or follow the shape of the particular piece to be forged. Dies with irregular parting lines are known as lock dies. The shape of the forging to be made also determines the steps of operation and the requirements for die design.

The die blocks are now ready for layout which is one of the important operations in the sinking of a set of drop forge dies. Before this can be done, however, the exact method of fabrication of the particular part must be determined. Fig. 8 shows one half of a forging die requiring fuller, edger, blocker, finished impression and

cutoff. Frequently there are designs requiring additional operations that must be considered when laying out dies. The layout upon each face is made with such reference to the matched edges that, when blocks are placed in the hammer with the edges in exact match, the layouts on each face are in exact coincidence.

The required impressions for a die as shown may be described as follows: The fullering operation reduces the side of the section of material used. The edging operation conforms the material to the area of the respective sections of the piece to be forged. The blocking operation gives the forging its general shape. The finishing operation completes the forging. The cutoff is used to cut the forging from the bar so the same operation may be repeated.

The sinking of the dies is the machining of the impressions and is usually done in a vertical milling machine, the die sinker following the outline of his layout. Upon the completion of the milling operation, there is the finishing operation, such as chipping, scraping, grinding, polishing and stenciling. After the impressions are completed, the dies are clamped together with the edges in exact match and lead is poured into the impression through an opening known as the gate or sprue, producing a model of the piece to be fabricated. This is used for checking purposes. All measurements in the layout, calipering of impressions and checking of lead models is done with a shrink rule which compensates for the shrinkage of the material. The amount of shrinkage allowed varies with the type and shape of forging and the grade of material used.

Upon completion of the dies for production, the shanks are machined for dowels which keep the dies from shifting when keyed into the hammer. The finish impression is then provided with a gutter into which the excess metal will flow during forging. The sharp edges at the parting of the impression are removed which provides an easy flow for excess material along the flash line and provides a space in the die to take care of the flashing. Upon completion of this the dies are ready for production.

In addition to the forging dies it is necessary to provide trimming dies or tools for the removal of the excess metal around the forging or flashing. These tools are known as trimming dies or tools and are usually furnished for either hot or cold trimming, depending upon the method desired to be used. Small forgings are usually cold trimmed. The trimming blades are made of tool steel and are shaped to fit the outline of the forging at the parting line

of the dies. The trimming punch is shaped to the contour of the top of the forging to be trimmed. Hot trimming is usually done in a press adjacent to the hammer at the time of the forging operation whereas cold trimming is usually done in a separate department, but likewise in a press.

There are many factors that enter into the manufacturing cost of dies and have a direct bearing on the life of a die. Some of these factors are tolerances, draft angles, radii and fillets. With very close tolerances, no matter how carefully the die is designed and handled in production or how good the grade of die material, the forgings soon spread beyond the specified die limits, due to die deformation, necessitating the re-working of the dies which is always an expensive operation. When the draft angles specified are too small there is a tendency for forgings to stick in the die impression, thus shortening the life of the die as well as retarding production. It is therefore of great importance that draft angles be made as liberal as possible. It has been found that seven degrees represents a standard draft angle for general use. Frequently deep holes in forgings give better results if a 10-degree draft can be used. All radii and fillets should be made as large as possible thus eliminating the possibility of cold shuts developing in the fabricating process, as well as adding strength to the forging at that point.

There are a number of important factors to be considered and controlled in the actual fabrication of the forging.

1. The heating of materials for forging.
2. The elimination of forging defects.

The details to be observed in the practice of heating for forging are controlled by six metallurgical factors:

1. The rate of heating.
2. The overheating and burning of the material.
3. The maximum forging temperatures.
4. The decarburization of the surface of materials.
5. Scaling.
6. The finishing forging temperatures.

The time of heating[3] should be just sufficient to bring the material to forging heat. For steel of 0.50 per cent carbon and less a minimum of five minutes per inch of diameter is sufficient for heat-

[3] "Practice for Heating Steel Bars for Forging," *Metals Handbook*, American Society for Metals.

ing bars up to three inches in diameter. Bars over three inches in diameter require a longer heating rate. For steels having higher carbon content than 0.50 per cent, somewhat slower rates of heating should be used. With some steels, especially alloy steels, preheating should be employed in order to avoid rupturing. It is advisable with larger sizes or irregular shapes to determine the rate of heating by imbedding thermocouples for the particular furnace and bar to be heated. To utilize the minimum of heating, it is necessary that the bars in the heating furnace do not touch, and that the furnace is of sufficient capacity.

Care should be taken to avoid the accumulation of oxide slag on the floor of the furnace, since the presence of such slag causes pitting of the bars or billets which come in contact with it. In oil fired furnaces, there is often an accumulation of coke formed at, or near, the burners, and occasionally other places. This coke, if not removed, may fall on the bars and billets, carburizing them at the point of contact and cause hard spots. Melting and pitting of the material also may result with some types of alloy steels, especially those containing chromium.[4]

Overheating is the result of prolonged heating at temperatures which produce excessive grain growth which can be restored by heat treatment or a combination of mechanical working and heat treatment.

Burning of steel consists of heating same to a sufficiently high temperature to produce abnormal grain growth and the destruction of intercrystalline cohesions which cannot be restored by a heat treatment or a combination of mechanical working and heat treatment.

The maximum forging temperatures recommended for safe practice is 200 degrees Fahr. below the solidus curve, while the minimum depends on the shape and size of the forging. Dr. W. H. Hatfield's suggested relative maximum forging temperatures are given[5] and have been found satisfactory:

	Degrees Fahr.		Degrees Fahr.
0.10 Carbon Steel	2400	0.90 Carbon Steel	2150
0.20 Carbon Steel	2375	1.00 Carbon Steel	2000
0.30 Carbon Steel	2350	1.50 Carbon Steel	1900
0.40 Carbon Steel	2325	Monel Metal	2100
0.50 Carbon Steel	2300	Copper	1800
0.60 Carbon Steel	2250	Brass	1400
0.70 Carbon Steel	2225		

[4] "A New Forging Defect, Its Cause and Cure," Jominy and Steever, *Metal Progress*, Jan. 1931, Vol. XIX, No. 1.

[5] "Practice for Heating Carbon Steel Bars for Forging and Forging Temperatures," *Metals Handbook*, American Society for Metals.

All precautions should be taken to avoid decarburization of the surface of forgings so that they will respond to heat treatment and thereby produce the physical properties expected.

Scaling of materials should be reduced to a minimum. It is the result of many forging difficulties. It is hammered into the surface of forgings, causing machining difficulties. Represents an economic loss of metal and loss of time in removal; it reduces the productive life of forging dies due to its abrasive action. As it is desirable to have as little scale as possible, a minimum time of heating should be employed.

The finishing temperature should be as near as possible to the critical range of the steel. This will produce a fine grain and enable the forgings to respond to heat treatment without difficulty.

Grain Flow

Proper grain flow[6] in the finished forging is of importance. In order to secure maximum strength and durability forgings should be so formed that the flow lines are in the correct position. This is true with forgings that are subjected to shock and high stresses. There are certain types of forgings that are of a simple nature and are not subjected to high stresses. Grain flow in these cases is not so extremely important. There are some types of forgings that involve considerable detail in fabrication to produce proper grain flow.

The lack of knowledge on the subject of grain flow is surprising, and the question of grain flow is seldom mentioned, because many engineers are not familiar with its importance.

The microscope reveals that the structure of steel is crystalline. By rolling or mechanical working the constituents are so aligned as to impart in the material a fibrous structure. This fibre can be detected by deep macro etching in suitable reagents.

The physical properties of steel, particularly impact, are greatly influenced by the existence of this fibrous condition giving it increased ability to withstand shock.

All metals are composed of crystals. Engineering materials are dependent upon strength, ductility, resilience, etc., thus the strength and general characteristics of the crystals, together with the nature and strength of the material which is present between the individual

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[6] "Fine Forgings Require Correct Heating and Grain Flow," Steever, *Metal Progress*, June 1932, Vol. 21, No. 6.

crystals, govern the final physical properties desired in the finished product.

All mechanical worked metals originate as castings in the form of ingots, and it is necessary to first consider the formation of ingots. Cast metals have no direction of grain flow and this applies to ingots. If an ingot be sectioned it is possible to observe its structure by suitable etching. From the surface all ingots appear much the same, but etching may reveal a variance sufficient to make it unsuitable for

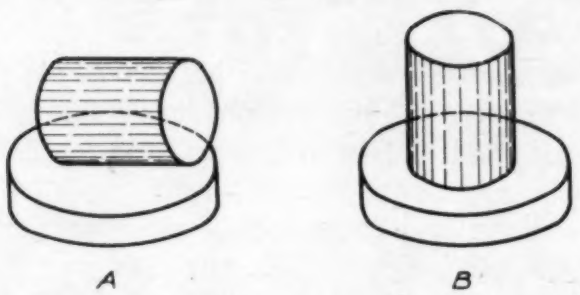

Fig. 10—Diagram Showing Forging of Bar Longitudinal and Upset.

sound rolled stock. When ingots cool the crystals separate in an unequal fashion usually arranging themselves in what is known as a dendritic structure. Such a structure can be compared to a fir tree appearance; a string of crystals form and shoot off branches which in turn shoot off other branches at right angles to them; this is repeated until the structure has an appearance similar to a pine tree. Although this form of crystallization takes place the important point to notice is that there is no definite flow. This can be proved by cutting test pieces and making physical tests. If such tensile tests are made from specimens taken from all directions it will be found that the results obtained are practically the same within experimental error, and excluding the effects of unsoundness.

The crystals of steel in the form of ingots are generally large and their size will depend upon the rate at which they are cooled from a specific temperature which varies somewhat in accordance with the composition of the steel.

The larger the size of all metallic crystals, other things being equal, the lesser is their adhesion and in the same degree their general mechanical properties. On the other hand the smaller the crystals of steel, the better is their adhesion and mechanical properties. It is evident, therefore, that small crystals are most desirable. It is for this reason that hot and cold working of materials is employed. The

effect of hot working is to break up the crystals and elongate them into long needle-like shapes. It is true when steel has been worked the direction of grain is very important and must be given due attention if the best possible properties and service are to be obtained from the material.

Physical tests made from certain types of forgings both parallel with and transverse to the direction of flow show interesting results. The most important difference in properties are the impact values.

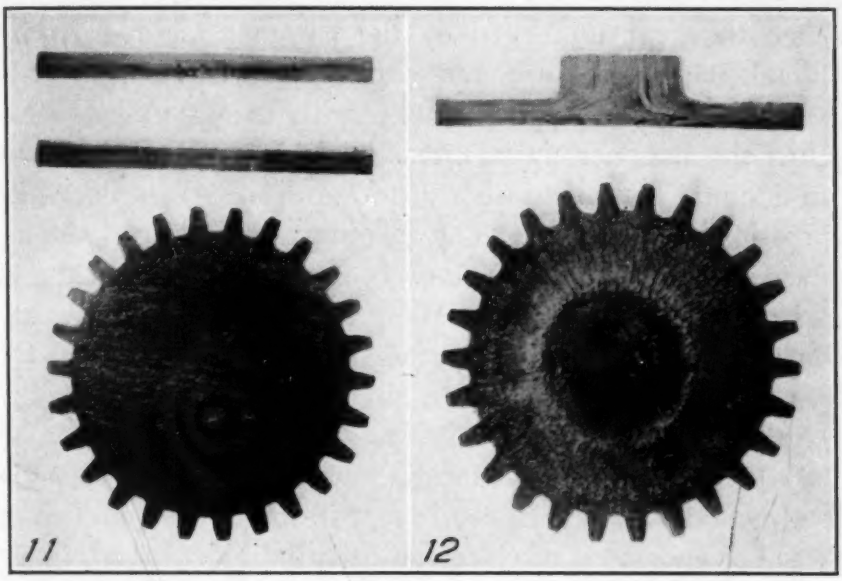

Fig. 11—Gear Forged Longitudinal.
Fig. 12—Gear Forged Upset.

The elastic limit and tensile strength are little different in these tests. The elongation and reduction of area are considerably lower on the transverse tests; these are important considerations because they govern the ductility of the materials.

Grain flow in steel follows the direction of rolling and runs parallel with its length; for example, consider a plank of wood in which the grain of the wood runs in the same manner. Supposing we take a plank of wood in which the grain runs transversely to instead of parallel with its length and it be supported at each end, it would without question, if it were long enough, fail, because of its own weight. If, on the other hand, the grain runs parallel to its length in the normal condition it possesses considerable strength and resistance to load. This example is to point out the importance of insuring, as far as possible, that stresses fall longitudinally with the direction of flow of the material.

Forgings possess a very distinctive direction of flow of their component crystals. Cast metals are comparatively weak and do not possess that characteristic. It is therefore important that the first essential in a forging is that it should not fail in service and also that stresses should only be allowed to fall on parts in such a direction as will be most resistant to failure.

Let us consider the forging of gear blanks. There are a number of ways to forge this particular part. Drop forge from a bar longitudinal. Drop forge by upset or slug method.

When the gear is forged by the hammer method from a bar longitudinal, it produces a blank with 50 per cent of the teeth cross grain and 50 per cent approximately with the grain longitudinal.

When the hammer upset method is employed, the slug of steel is set on end and flattened into a disk, the disk is then placed into the die impression and finished. The forging machine grips the material and by several passes or operations forms the blank into the desired shape. In the last two methods the grain of the material takes up a direction which is approximately radial from its center and the gear possesses a grain which has an appearance of a spoked wheel. These grains run along each individual tooth, and fracture is much less likely to happen than in the case where the grain does not take up this direction. When a gear is run in mesh with another gear it is subjected to enormous stresses and each individual tooth must be as strong as possible, for the failure of one tooth is as serious as the failure of several. If the gear possesses cross grain at the root of the teeth, excessive stresses will probably cause failure, for several reasons, one being that the impact values taken transversely are low, and the other that a tooth with a transverse grain must withstand all the stresses, whereas, when the grain is longitudinal, part of the stress is borne by the center portion, owing to the grain in the tooth being connected with that in the center of the gear.

Hammer Forging

An outline of forging equipment, dies, heating and grain flow has been given and applies to both drop forging and hammer forging. Drop forging might be termed a metallic shape formed by hammering a cold or heated bar or billet between a pair of dies having the desired impression of the forging shape. Hammer forging might be termed a metallic shape produced by hammering, the shape de-

pending upon the skill of a smith. In the production of hammer forgings a large range of tools is needed. Space does not permit a description of the various types of tools in detail, but the following list is given to show those needed for forging rounds, squares, flats and octagons: Cutters, swages, pegs, tongs, measuring bars and tools.

Forging Rounds—The billet is cut to approximate weight, heated

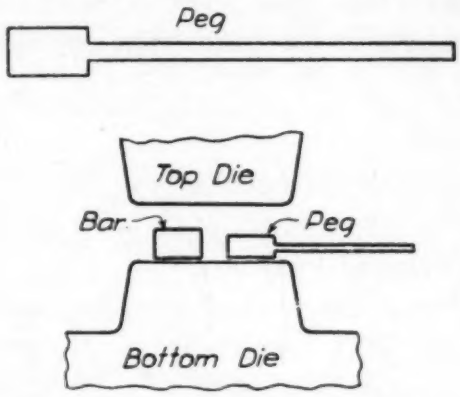

Fig. 13—Peg Used in Hammer Forging.

and forged to approximate dimension on all sides as a square. The corners are then broken and the billet has some resemblance of a

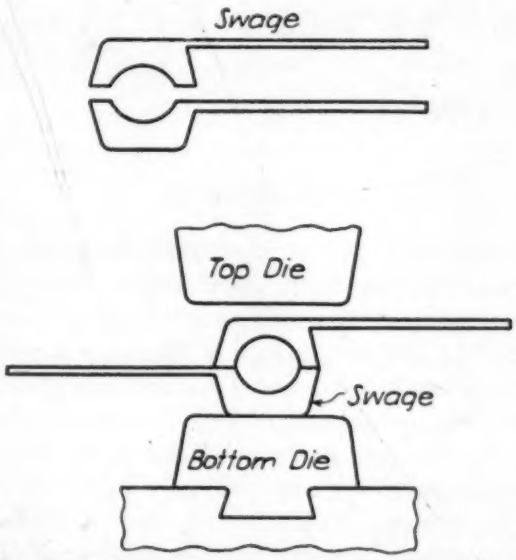

Fig. 14—Swage Used in Hammer Forging.

round. It is then placed between the swaging dies and finished into a round. The swaging dies are made to size with allowance for shrinkage.

Forging Squares and Flats—The billet is cut to approximate weight, heated and forged to size. In the case of flats, two pegs are used, one for the width and one for the thickness. The peg is placed upon the bottom die next to the bar being forged; the top die striking the peg governs the thickness of the bar being forged. The peg is made to proper size and allowance made for shrinkage. In forging squares or octagons only one peg is required.

The forging of rings and other irregular shapes is done in much the same manner with the addition of special tools, punches, etc.

Conclusion

The author has covered the practical side of forging manufacture and has endeavored to bring out a few of the many problems encountered in their manufacture.

The successful manufacture of forgings is the result of close co-operation between steel manufacturer, designer, machinist and forger. With continued co-operation in these industries, we can, without question, manufacture a superior metallic shape. This will please all concerned.

The author is of the opinion that the forging industry has contributed largely to the success of the automotive and other allied industries; also, that it is one of America's greatest assets for the production of munitions for national defense and, as such, has a secure future for continued success.

DISCUSSION

Written Discussion: By W. A. Spindler, assistant professor of metal processing, University of Michigan, Ann Arbor, Michigan.

Covering so broadly the many factors involved in the production of drop and hammer forgings, the author has had to discuss some of these factors quite briefly. This briefness applies especially to statements made with regard to finishing temperature, which the author disposes of as follows: "the minimum forging or finishing temperature depends on the shape and size of the forging and should be as near as possible to the critical range of the steel." The writer believes that, with the present availability of steels with high coarsening temperatures in both carbon and alloy grades, the subject of finishing temperature can profitably be considered at greater length.

Epstein on two occasions[1, 2] has suggested that steels with high coarsen-

[1]M. A. Grossmann, "On Grain Size and Grain Growth," Transactions, American Society for Steel Treating, Vol. 21, 1933, p. 1079-1104. Discussion by S. Epstein.

[2]S. Epstein, J. H. Nead, and T. S. Washburn, "Grain Size Control of Open-Hearth Carbon Steels," Transactions, American Society for Metals, Vol. 22, 1934, p. 942-973.

ing temperatures may be utilized to produce small "actual" grain size in the "as rolled or forged" condition, thus eliminating the necessity for normalizing. This would be done by restricting the heating for hot working to a temperature below the coarsening temperature which, in some cases, is as high as 2150 degrees Fahr. Inasmuch as hot working reduces the austenitic grain size, the writer believes that in many instances the stock for forging could be heated above the coarsening temperature and still produce an actually fine grain forging if the finishing temperature were below the coarsening temperature of the reduced section. This result would be dependent upon sufficient working of all parts of the forging for grain refinement. With forgings of varying cross section, this would not always be feasible so that, in these cases, the heating temperature for forging would have to be kept below the coarsening temperature of the bar stock or billets. The amount of reduction for complete grain refinement and the effect of reduction on the coarsening temperature, therefore, are important factors.

The effect of reduction of section by forging on the coarsening temperature has been studied by Grossmann,[3] who found that, in general, the coarsening temperature was reduced with increased reduction. His findings were based on subsequent carburizing tests, but should in general apply to "as forged" structures as well, as indicated by some preliminary tests made by the writer. Supporting evidence also is found in a paper by Grove.[4]

The above considerations lead to a tentative restatement of the optimum finishing temperature, that is, the finishing temperature for small grain size should be just below the coarsening temperature, for short time heating, of the smallest section of the forging. With the current interest in grain size control of steel, further information along this line relating to the production of forgings should soon be forthcoming.

Written Discussion: By J. A. Succop, Heppenstall Co., Pittsburgh.

The author, Mr. Adam Steever, is to be congratulated for giving us a very concise, clear and instructive outline of the drop hammer practice and its development. The author has covered an enormous field and despite the brevity in which he has handled the description of the practice and the necessary equipment, there remains little of importance to be added.

It is apparent in the picture drawn for us by the author that to fully realize the inherent possibilities of the drop hammer, it is necessary to be fully informed on the various components making up the hammer unit. It is difficult to point to any particular part as especially representative of the greatest progress since in all members there have been notable advances. The improvement of the complete hammer, with its more rigid frame and forged tie plate, with greater steam efficiency and quicker action, has materially assisted in the increasing of production of drop forgings.

A very apparent component, as you will instantly appreciate, is the die block, which part is being constantly replaced due to either failure by wearing out of the impression, or to a change in die design. The progress made in im-

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[3]M. A. Grossmann, "Grain-Size in Metals, With Special Reference to Grain Growth in Austenite," TRANSACTIONS, American Society for Metals, Vol. 22, 1934, p. 861-877.

[4]H. A. Grove, "Relation of Hot Working to McQuaid-Ehn Grain Size," Preprint Number 14 for A. S. M. annual meeting, 1935.

proving this particular part very clearly indicates the accompanying progress that has been made in engineering and metallurgy and attaches added importance to the fact that where there is sufficient stimulus for improvement, there will be adequate response.

From the plain carbon steel, longitudinally forged, unheat treated block, evolved the special alloy upset forged and hardened and tempered block of today. The most generally accepted analysis of today's highest quality drop forging die block is nickel-chromium-molybdenum. This chemical composition has proven to possess those properties required in a die block, namely, resistance to impact, abrasion and checking at elevated temperatures and minimum of distortion at high hardnesses.

It is necessary, in order to secure impact resistance in both the longitudinal and transverse direction, to upset forge the die block during the press forging operation, distorting the original ingot fibers so that the completion of the several upsetting operations, the die steel contains a minimum of directional properties. When the die is in service, the impact stresses are absorbed by the die block in all directions and therefore only a quality block can successfully withstand these stresses equally both transversely as well as longitudinally. In the early history of the manufacture of drop forging die blocks, the consumer bought dies in the annealed condition and after machining the impression, the dies were hardened and tempered. This necessitated a reannealing and rehardening of the die each time a new impression was machined. Since the introduction of alloy steels and since the modern development of the machine tool and the improvement of the technique of machining hardened steel, the sale of hardened and tempered die blocks hardened for specific service, has predominated. Such blocks are hardened to a specific hardness for a definite service and need not be reheat treated during the life of the block, regardless of the number of resinkings of the impression.

It is a safe conjecture that ninety per cent of the tonnage of all the die blocks in service today are purchased in the hardened and tempered condition, and the annealed type, hardened by the customer, is sold only for very shallow impressions and then the service requires a hardness above the commercial machinable hardness which is normally accepted as 2.95 Brinell ball diameter, 429 Brinell hardness number or 60 scleroscopic hardness.

The ability of the modern die sinking machine shop to machine dies with a hardness of 60 or even 62 scleroscopic hardness, is worthy of mention, for it was only a few years ago that the maximum machinable hardness of dies was considered 56 scleroscopic hardness. The type of impression generally governs the required hardness and a guide for the selection of hardness is given in the following table:

Hardness Grade	Brinell Ball Diameter	Scleroscope	Depth of Impression	Average Cross-Section Size of Block
A	2.95-3.05	60-56	½" max.	16 × 10
B	3.10-3.20	55-52	1" max.	18 × 14
C	3.25-3.35	51-48	3" max.	27 × 16
D	3.40-3.55	47-44	Unlimited	Unlimited

In addition to the liquid quenched and tempered die block, there is available for service still another type of block which is the normalized and tempered

type. This treatment is generally applied to die blocks which are extremely massive and are to be used in the drop forging of very large and heavy sections.

The dies of today have a wide variation in size running from 3-inch cubes, weighing approximately 8 pounds, and are used to produce tools and other very small sections to 36 by 24 by 56 inches long, dies weighing 13,000 pounds, used to forge rear and truck housings, diesel crank shafts and other heavy sections.

A most recent development in the perfection of the hardened and tempered die block is the tempering of the shank of "A" type of block which is the die block of the maximum machinable hardness. The shank is tempered to a hardness approximately 8 to 10 points on the scleroscope below the face hardness of the die. This softening of the shank side of the high hardness dies assists the die sinker to more easily machine the dovetail of the shank and materially increase the impact resistance in that portion of the die where high impact strength is most needed. Blocks having a hardness below the hardness of the "A" type do not need a softer shank due to the steel being readily machinable and possessing a sufficiently high impact resistance.

The improvement in the ram and sow block used in drop hammers is also somewhat notable. These parts are also susceptible to impact failures and an effort has been made by the various manufacturers to strengthen these essential parts. Sow blocks and rams weighing up to 16,000 pounds are being made of alloy steel and after machining are oil-quenched and tempered to obtain physical values which enables the part to withstand the extreme abuse in service. The greatest stresses in a sow block and a ram are developed parallel to the die. seat and by a newly adopted method of cross forging, the original longitudinal direction of the ingot as cast is made to run transversely at mid-length of the sow and at mid-length of the height of the ram between guide slots. This places the original center of the ingot at locations in the sow block and the ram where the least stresses are developed. The die block is keyed into position in the ram or sow block, at right angles to the original direction of the ingot, thereby materially strengthening the die slot.

Steam hammer piston rods are another weak link in the service of a drop hammer. In the writer's opinion, and based on the experience of many hundred failed rods, we can readily state that the vast majority of failures has been due to mechanical causes, rather than to the quality of the metal or to poor heat treatment. The trend in recent months is to raise the hardness of the piston rods, materially raising the elastic limit at a sacrifice of the ductility, thereby increasing the endurance fatigue life of the rod. Regardless of the elastic limit or the ductility of the steel, the mechanical fit between the tapered end of the rod and the bore of the ram becomes the governing factor of the life of that particular rod. Of course, the quality of the steel and the subsequent fabrication will always play an important part and materially assist in prolonging life in service.

A modern alloy steel piston rod after having been forged from a proper size ingot and normalized, quenched and tempered, should show approximately the following physical properties:

Ultimate Strength	140,000 pounds per square inch
Yield Point	120,000 pounds per square inch
Elongation in 2 inch	20.0 per cent
Reduction in Area	55.0 per cent
Brinell Hardness	300 Brinell hardness number
Izod Impact Value	50 foot-pounds

Obviously, the aforementioned members included among the many in the drop hammer unit, necessitate careful consideration of quality, not only chemical and physical, but mechanical as well, that will enable them to stand the abuse and punishment of the drop forging operation and still keep on hammering out peak production.

Written Discussion: By W. E. Jominy, Research Laboratories Section, General Motors Corporation, Detroit.

This paper covers a very wide scope and has in it much useful information. In my judgment the suggestions for heating for forging are sound. As the author states, the maximum forging temperatures recommended are well below the temperatures at which burning occurs. Burning is to be avoided at all costs and it is well to be cautious about heating and forging at too high a temperature. Not only is temperature control necessary but it is desirable to be careful of the atmosphere when higher temperatures are used. A jet of air impinging on hot steel will cause it to burn even though the furnace temperature is considerably below the solidus curve of the iron-carbon diagram. On the other hand with good control of temperature and atmosphere it is possible to use higher temperatures than those given. The safety of such procedure depends on how accurately the temperature of the steel in the furnace can be controlled. It scarcely seems necessary to add that where no pyrometric control whatever is used it is better to stay under the temperatures given in this paper, and yet it is in such installations that high temperatures are often used.

Quite commonly sufficient room is not available in the forge shop for furnaces of such size that slow heating to a proper forging temperature is possible. The result is undersized, overheated furnaces in which the steel can be heated very rapidly but in which the steel cannot be permitted to reach the furnace temperature if danger of burning is to be avoided. Such furnaces are usually the slot type in which the back of the furnace is much hotter than the front. Using this type of furnace we are entirely dependent upon the man in charge of heating the steel for safe forgings.

When such furnaces are used, and they are very common, it would seem advisable to hold the maximum furnace temperature a little below the burning temperature of the steel. This would still permit fast heating and at the same time safeguard against burning the forging. This will require temperature control equipment which is accurate at high temperatures and while such equipment is not cheap it does seem worth while.

The author states that "the finishing temperature should be as near as possible to the critical range of the steel. This will produce a fine grain and enable the forgings to respond to heat treatment without difficulty." While the second sentence in this statement is undoubtedly true, I wonder whether the author would apply the first sentence in this statement to alloy steels. Is it not customary in drop forging alloy steels to work to a finishing temperature not lower than about 1800 degrees Fahr.?

Author's Closure

Many interesting and important points have been brought out in the preceding written discussions. Mr. Succop contributes a valuable addition regarding dies used in forging. Professor Spindler contributes a valuable addition with regard to finishing temperatures. Mr. Jominy's remarks on the subject of maximum forging temperatures and factors relating to heating contribute much additional information on this important subject. In reply to Mr. Jominy's question, "Is it not customary in drop forging alloy steels to work to a finishing temperature not lower than about 1800 degrees Fahr.?" the author's statement that "the finishing temperature should be as near as possible to the critical range of the steel" applies to both alloy and carbon structural steels, as well as alloy and carbon tool steels. Although alloy steels require more work in terms of impact at the low finishing temperatures, which in turn shortens die life to some degree, depending upon the amount of finishing per forging, tests show that the improvement in structure compensates for the slight loss in die life.

HOT PRESS AND UPSET FORGING

By J. H. Friedman

Abstract

Press forging and upset or machine forging are essentially one and the same process. While the equipment used is totally different in appearance, the press and forging machine perform their work in the same manner, i.e., by preliminary or gathering operations and final or finishing operations. The principal difference between the two methods is that the forging press generally employs solid dies, whereas the forging machine uses split dies. Also, the hot stock for the press is generally in the form of cut blanks or slugs, while bar stock is used in the forging machine. It is true that split dies and bar stock are sometimes used in the forging press, and solid dies and cut blanks in the forging machine, but such applications are the exception rather than the rule. Due to the similarity of the two methods, they could be dealt with together as one subject. However, for purposes of this symposium, they will be handled separately—the forging press first, and the forging machine second.

Forging Press

Types and Application

MANY types of presses such as hydraulic, friction-screw, toggle, crank, etc., have been tried on hot forging work with very indifferent success. These early presses were developed principally for drawing, stamping and pressing operations on cold material; and, when hot work was attempted, it was found that die life was so poor as to make the process impractical from an economic standpoint. Consequently, rather than discuss the various types of presses, it will be more pertinent and of greater interest to cover the essential features of a modern hot forging press, and point out why these features are desirable on this type of work. It is a coincidence, however, that the most successful type of hot forging press on the market today was

A paper presented as part of the Symposium on the Plastic Working of Metals, Eighteenth Annual Convention of the Society held in Cleveland, October 19 to 23, 1936. The author, J. H. Friedman, is vice president, The National Machinery Co., Tiffin, Ohio. Manuscript received June 25, 1936.

also developed for a class of work other than hot forging. For this reason it is necessary to digress for a moment to explain how this modern press came into being.

With the advent of the automobile and the mass production era, the hammer and forging machine received first attention from designers and development engineers, because these machines had the widest application and were in the greatest demand. The ever increasing need for better and more productive forging machines to handle the constantly mounting varieties of alloy steels and to keep pace with production requirements, resulted in rapid and revolutionary improvements. These new and modern machines make possible the mass production of low weight, minimum finish forgings, and, where holes are required, they are pierced to size ready for broaching.

As developments in the forging machine progressed, the many savings that were being effected in material, heating, handling and machining costs, made it obvious that still further savings would result by pressure finishing the machine forgings by a coin forging operation, thus bringing them still closer to the desired finished size and reducing subsequent machining and scrap loss. The micrometer caliper had displaced the rule as a measuring instrument. As no product can be more accurate than the machine in which it is made, designers realized that to meet the requirements of hot coining a radically different type of press would have to be devised. As a result of much development work along this line, the modern coining press was introduced. This is a "stallable" press with a massive, one-piece, heavily crowned, heat treated steel frame, pre-loaded with four large shrunk-in tie bars. The main shaft is of full eccentric type, the pitman is solid and connected directly to the ram. The ram is of the over-arm type, the knock-out is rugged and positive, and the machine is fast and convenient to feed. As a result of the combination of these features, "spring" has been reduced to the vanishing point, die alignment is extremely accurate, ejection is quick and positive and production is greatly increased. A press of this type is shown in Fig. 1.

These same features of extreme rigidity, accurate die alignment, positive ejection, speed and convenience of feeding are also essential in a forging press for reasons which will be pointed out later on. It will be seen, consequently, that the development of the modern coining press also resulted in a machine that is ideal for hot forging work.

Experience has proven conclusively that machine rigidity has a major effect upon metal flow. When a forging is produced in a press having a weak bed frame, which stretches under load during the final stage of the forging operation, the stock, following the path of least resistance, flows out between the faces of the dies in the form

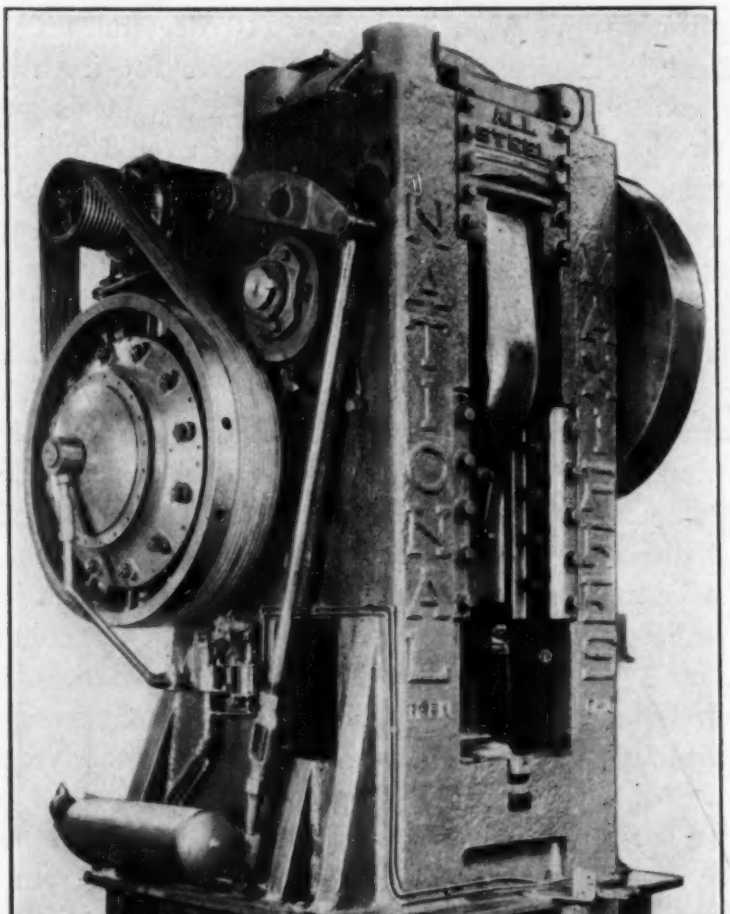

Fig. 1—Photograph of Modern Coining Press.

of a flash rather than into the die recesses. The result is a forging with an excessive amount of flash; and, at the same time, one which is not properly filled. The more complicated the die the less the chance of filling the forging. Setting the dies closer does not solve the problem; but merely results in additional "springing" of the bed frame; and, if carried far enough, breakage occurs.

The importance of machine rigidity and its effect upon metal flow was not fully appreciated until the demand for bolt headers and forging machines to handle alloy steels brought improvement in bed frame rigidity on that type of equipment. It is true that there are not many opportunities to observe the effect of machine rigidity on

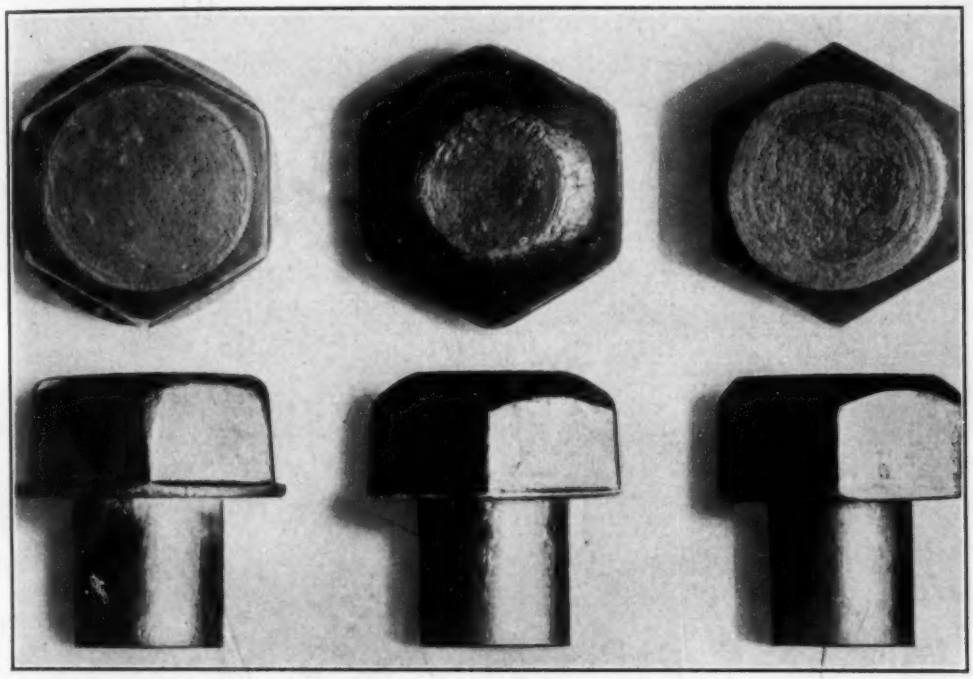

Fig. 2 Fig. 3 Fig. 4

Fig. 2—Shows a Hexagon Head Bolt Made in One Blow on the Older Type of Heading Machine.
Fig. 3—Shows the First Stage of a Similar Bolt Made in One of the New Style Machines.
Fig. 4—Shows Restruck Bolt After Operation Shown in Fig. 3. Note Sharp Corners and Parallel Sides.

metal flow, because each job presents so many variables that comparisons are not easy. However, in the manufacture of bolts the large quantities produced combined with the uniformity of controlling factors give one the best opportunity to make such observations.

Fig. 2 shows a hexagon head bolt made in one blow on the older type of heading machine, in which it will be readily seen that the top corners of the bolt are not well filled even though a large flash has been thrown at the base of the head.

Fig. 3 shows the first stage of a similar bolt made in one of the new style, more rigid bolt forging machines in which the head is also formed in a cupped heading tool, just as in the case of Fig. 2. It will be noted, however, that the sample shown in Fig. 3 is well filled

and has sharp corners at the top of the head with but a small flash or overflow at the base of the head.

It is customary to strike such bolts an additional blow in another impression of the die, in order to remove the draft on the sides of the head and press in the flash at the base of the head; and the finished bolt, after this operation, is shown in Fig. 4. The interesting part, however, is that the first or cupping operation, as shown in Fig. 3, when made upon the new, more rigid forging machine has sharper corners at the top and less fin at the base than the bolt shown in Fig. 2, made in an old style machine. It is obvious that all this is the sole result of machine rigidity, as all other factors, such as speed of operation, heats, etc., remained constant.

If we assume that the improved quality of the bolts made upon the new style, more rigid machine resulted from the increased stiffness of the machine, then it logically follows that this added stiffness also had an effect upon the velocity of metal flow. The velocity of flow would certainly be better maintained in the new, stiffer machine, than in the old one. In the old machine, when the crankshaft approached its forward dead center, the weaker frame and working parts stretched under the load when the heading pressures started to mount. As a result, the movement or velocity of metal flow at that time drops perceptibly on account of the considerable amount of spring in this old style machine during the last part of the forward movement of the shaft. Very little seems to be known about the subject of the effect of a change in velocity of metal flow in forging, and very little information is available in regard to it. However, there is considerable evidence in connection with the operation of drawing wire and similar work, and on hot and cold extrusion, which indicates that satisfactory results are obtained only so long as there is no appreciable reduction in working velocity. Evidently, once the metal starts to move, no trouble occurs if it is kept in rapid motion, but should the operation pause for any reason, it is generally found impossible to cause it to again start to flow. Possibly when a given velocity of flow is reached, a degree of molecular heat is maintained which assists in the metal flow. It has been observed that the gage on a hydraulic extruding press might drop from 300 tons at the beginning of an operation to 195 tons when the ram reached the bottom or the most difficult part of its stroke, which would seem to be a paradox; but, if for any reason the movement of the ram was momentarily stopped, the full capacity of the 500-ton press would not again

start it. The point is, there is a great amount of evidence to prove the desirability of stiffness in equipment of this type, and none to disprove it.

Granting that rigidity is essential in a good forging press, probably the next most important factor is speed. This matter of speed should not be confined to a fast revolving flywheel or a rapidly reciprocating ram. It should be considered in a much broader sense. Convenience and ease of feeding, quick starting, positive and rapid ejection, etc., are all vitally important factors of ultimate speed, and have their effect upon not only output alone, but die life as well. There is a constant flow of hot stock passing into the dies in a forging press which tends to draw the temper prematurely. Consequently, if satisfactory die life is to be obtained, it is essential that contact of the hot stock with the dies be as short as possible. A forging press that meets these requirements is convenient to feed, it starts the instant the control pedal is depressed, the ram completes its stroke rapidly, and the instant it starts on its back stroke the knock-out raises the forging out of contact with the die.

Summing up this matter of the essential features of a good forging press, it must be speedy, the entire machine must be rigid and free from spring, the die alignment must be accurate, the power of the dies should be ample and positive, it should be convenient and easy to feed, it should have instantaneous starting, and it must be provided with rugged and accurately timed knock-outs.

The application of the forging press on steel forgings is confined to comparatively simple and symmetrical forgings. It should not be recommended for preliminary breakdown or drawing operations as such work can be handled to the best advantage on equipment having an adjustable stroke or blow, such as a drop or a Bradley type hammer. The press is also used to advantage for coin forging at low temperatures (below the scaling point), blanks previously prepared on the hammer or forging machine. Typical jobs of this classification are shown in Figs. 5 and 6. This forging, Fig. 5, was trimmed and then normalized, the pressure finishing being done on the normalizing heat. The upper view shows the secondary flash or excess volume. Welded, hammered or upset blanks, like that shown at the left, Fig. 6, are press-forged accurately to size. Rough machining is eliminated.

Forgings that have projecting lugs, undercuts, etc., which require an opening die to free them, are, in the majority of cases, best handled

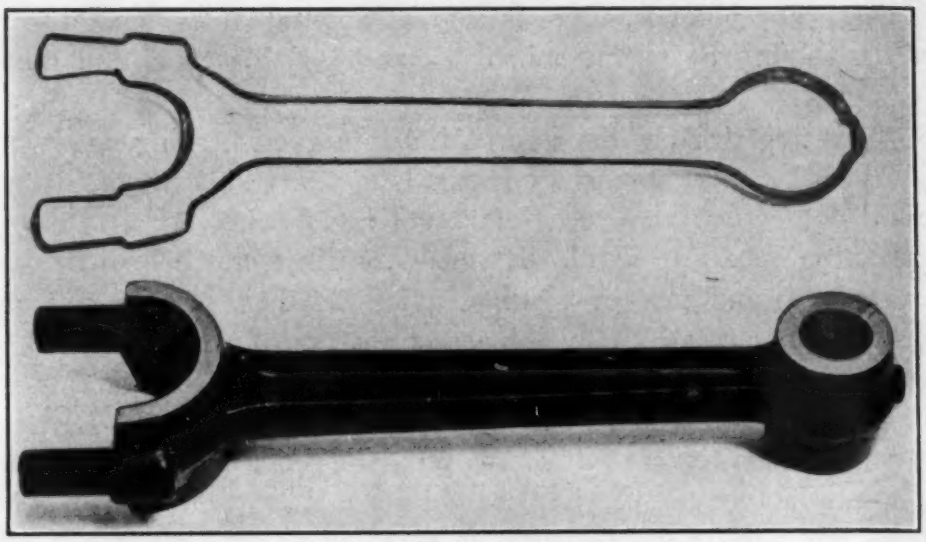

Fig. 5—Connecting Rod Coined on a Modern Forge Press on Normalizing Heat Showing Secondary Flash Trimming.

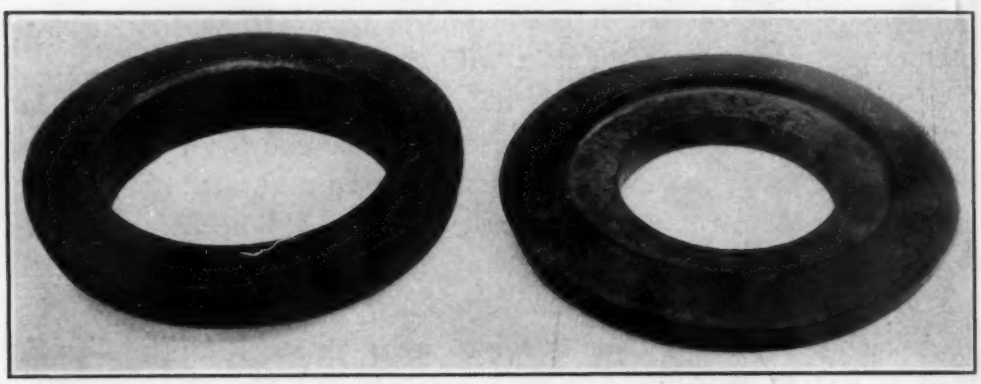

Fig. 6—Welded, Hammered or Upset Blanks, Like That Shown at the Left, are Pressed Accurately to Size. Rough Machining is Eliminated.

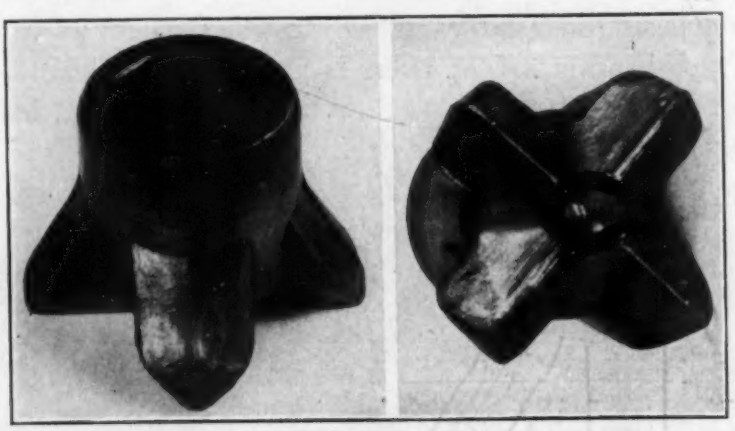

Fig. 7—Rock Drill Bit Made on a Modern Forging Press Using an Open Type of Die.

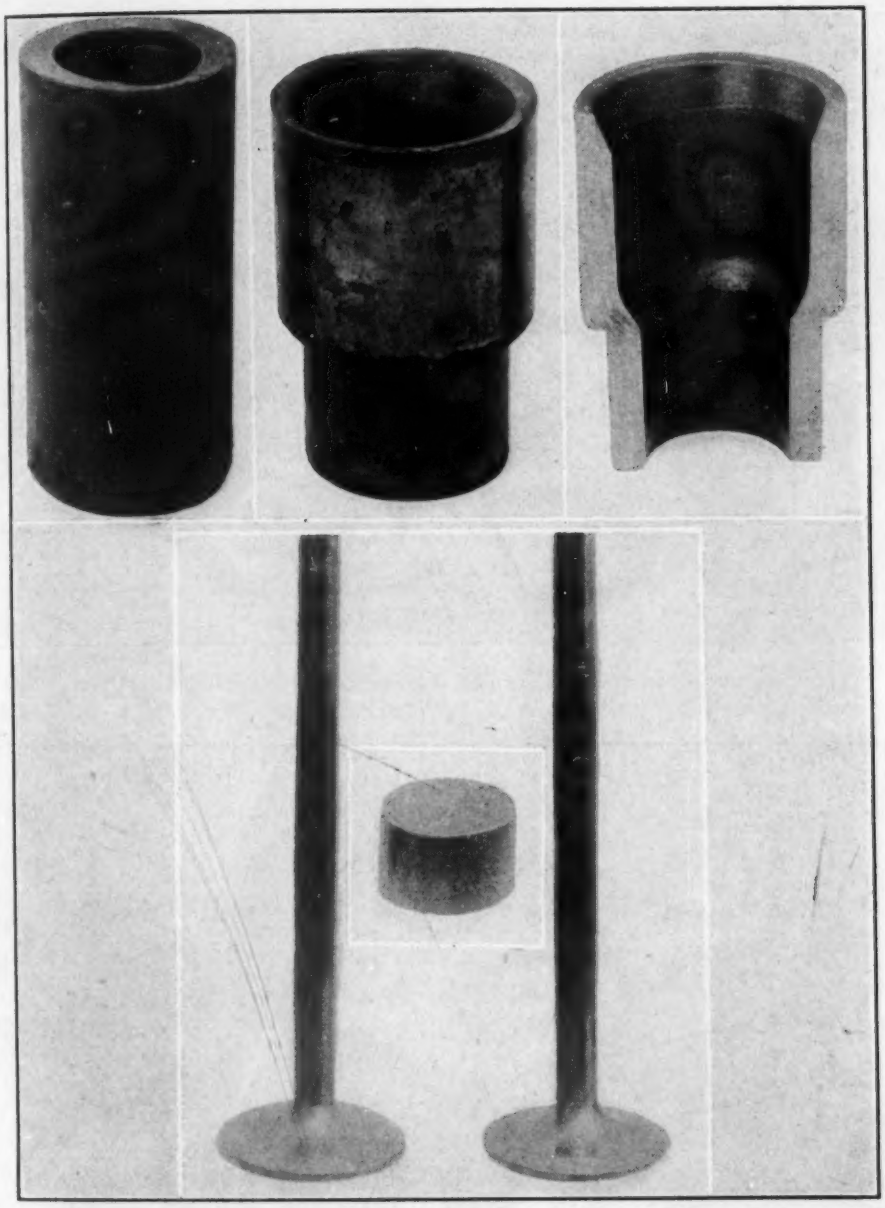

Fig. 8—Upper Group, Showing a Forging Pressed From a Tubular Blank.
Fig. 9—Lower Group, Shows Valves Extruded From Small Blank as Shown in Center.

on the forging machine. There are exceptions, of course, one of which is the rock drill bit shown in Fig. 7. Here is a forging which must be made in an opening type of die, but, due to the fact that it is a cut blank job, it can best be handled on a press.

A different application is shown in Fig. 8. This part is press-forged semi-hot from the tubular blank shown at the left. Due to the use of a solid die, there are no part line fins, and the forging is held

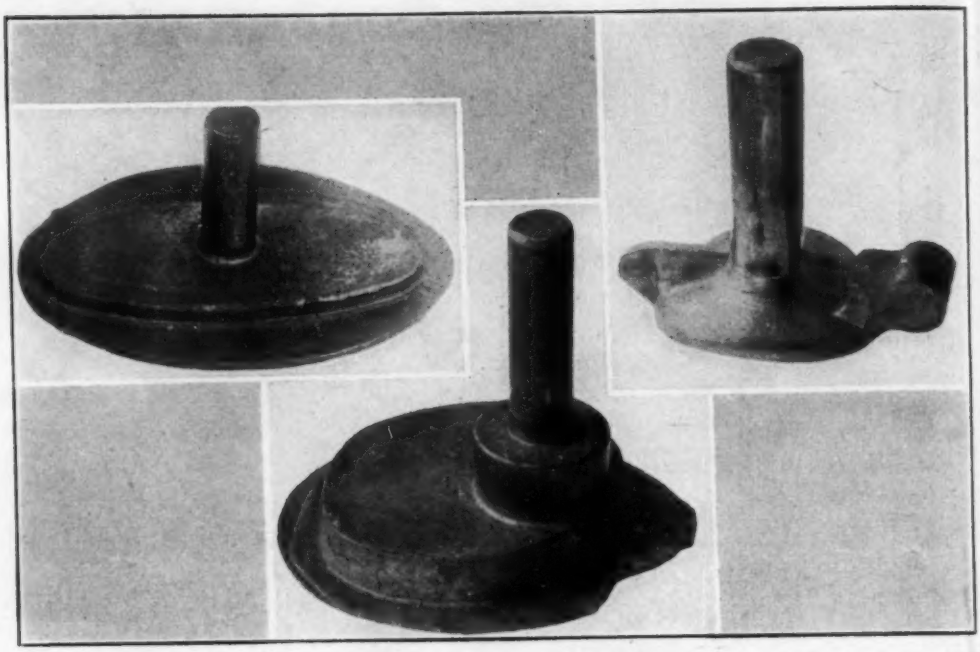

Fig. 10—Photograph of Typical Forgings of the Extruded Type.

Fig. 11—Showing Group of Symmetrical Forgings Which are Ideal for the Modern Forging Press.

within grinding limits. Fig. 9 shows a valve which is extruded from a small blank, as shown in the center. Two operations complete the forging ready for grinding. The stems of the forgings shown in Fig. 10 are extruded. Automatic ejection enables the elimination of draft, which, in turn, eliminates much machining time and scrap loss.

Symmetrical forgings, such as those shown in Fig. 11, are ideal for the press. They are all forged from cut blanks. Most of them are three operation jobs, the first blow being a breakdown operation to crackle off the scale, the second a partial forming operation, and the third a finishing operation. All are free from undercuts; consequently, can be made in solid dies. Due to the use of solid dies, there are no part line fins; and mechanical ejection eliminates the necessity for draft. These are ideal examples of low-weight, minimum-finish press forgings.

These illustrations show not only the type of forging work that is being handled in an entirely satisfactory manner on forging presses; but also give an idea of some of the other types of jobs which have been tried with more or less success. No doubt further improvements in dies and die steels and the use of the insert type of dies will greatly add to the class of work that can be handled to advantage on the forging press. New jobs are constantly turning up that are developed into satisfactory press work.

While it will be seen from the foregoing that there are certain limitations to the successful application of the forging press on steel, the reverse is true in the case of nonferrous forgings, as press application in this field is almost unlimited. The nonferrous metals are forged at such a low temperature that the heat has no serious effect on the dies, and most of the nonferrous alloys can be easily extruded or squirted into intricate die recesses. Also, nonferrous work produced on the press has an extremely good finish, in fact, just as accurate and smooth as the die itself. Consequently, parts that are to be plated require much less work to prepare them for the plating operation when they are produced on the forging press. Naturally, the same features that are desirable in a press for steel forging make it just that much better on nonferrous work, and experience has definitely proven that the forging press is in a class by itself for the mass duplication of nonferrous parts.

Flow of Metal Fiber and Grain Size

There is no fundamental difference between press and machine

forging as regards control of fiber flow lines, grain size for density and warpage and the temperatures used in forging. Consequently, these points will be discussed later on, under machine forging.

Machine Forging

Types and Application of Equipment

In addition to the standard, or high duty, there are a number of special forging machines. Among the machines which may be clas-

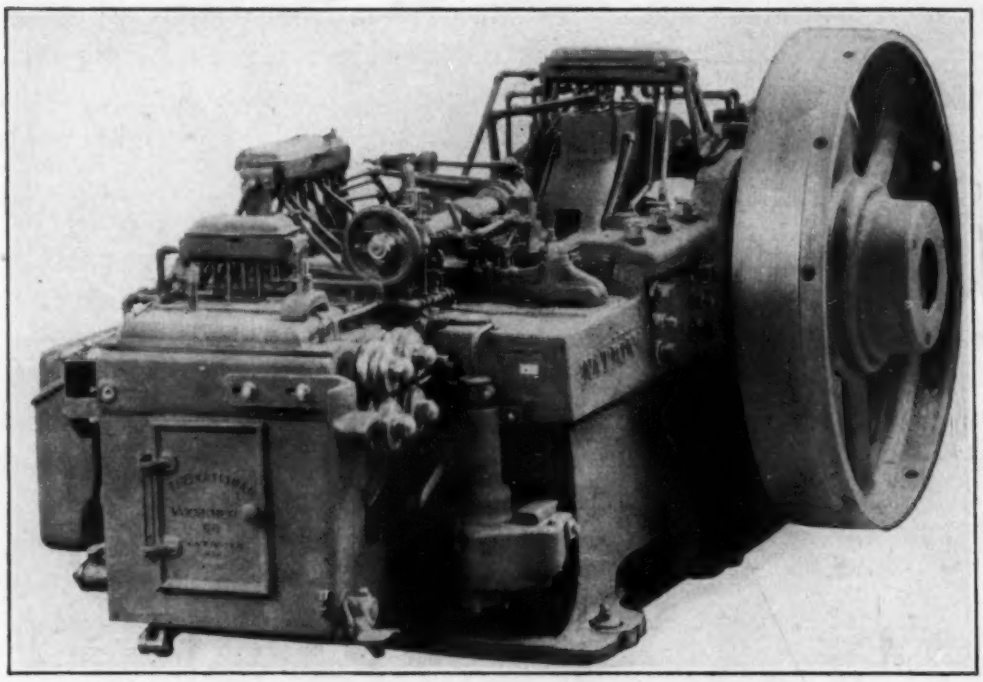

Fig. 12—Photograph of Continuous Motion Automatic Feed Bolt and Rivet Forging Machine.

sified as special, are bolt and rivet headers, ball headers, spike headers, and electric headers.

In order to conserve space, these special machines will be discussed only briefly—the most of this portion of the subject being devoted to the high duty forging machine because of its more general interest and wider application. As the stop motion bolt header is, in reality, a small high duty type forging machine, it will not be discussed separately.

Rivet forging machines are built in continuous motion hand feed and continuous motion automatic feed types. They are used for making rivets and bolts, and are confined to work that can be made in one blow.

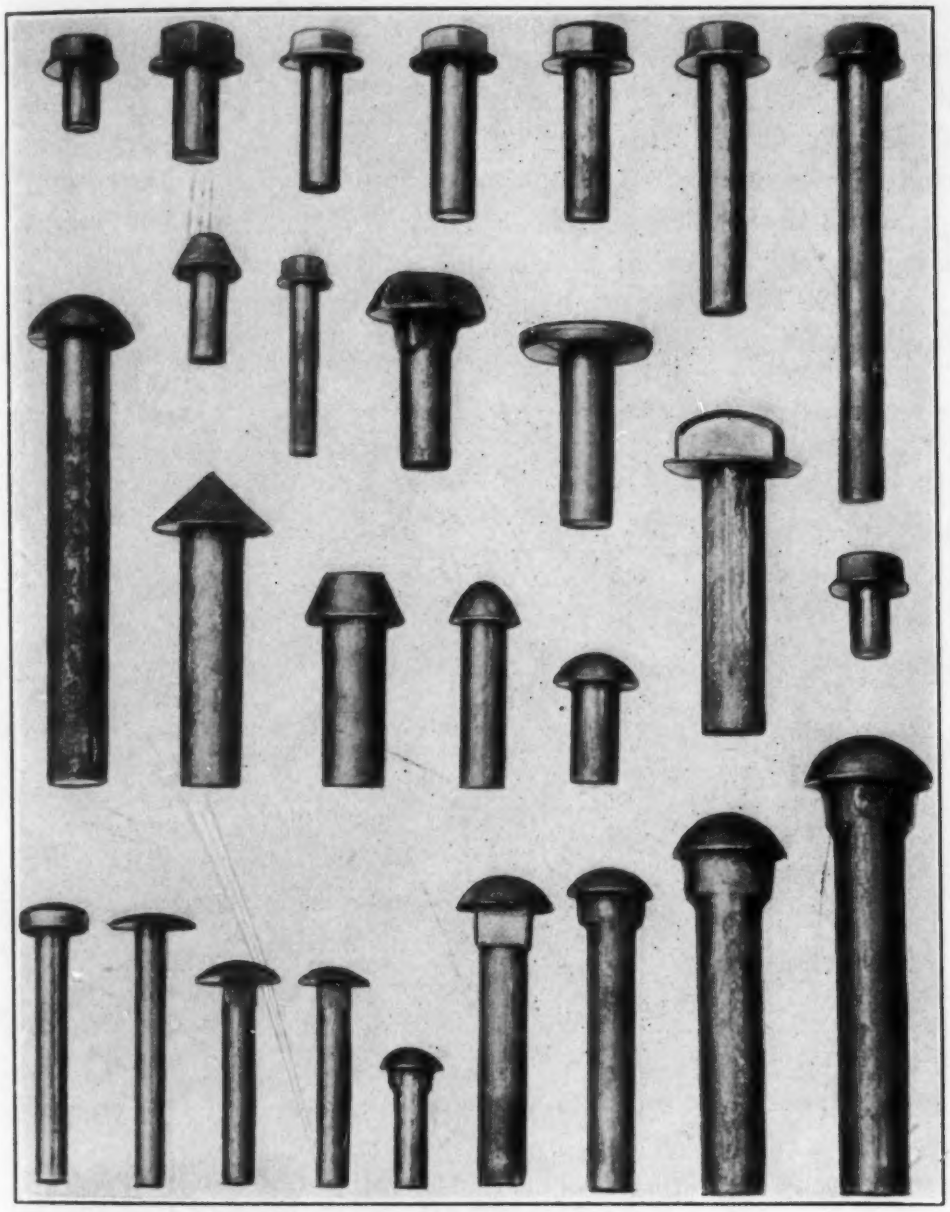

Fig. 13—Photograph of Some of the Products of the Forging Machine Shown in Fig. 12.

In case of the hand feed machines, the stock is in the form of bars or rods six to eight feet in length. The bars are heated for a distance of two or three feet and are fed into the machine until the heated portion is used up, then returned to the furnace for another heat. This is repeated until only a short crop end remains. The machine cuts off a blank of proper length, the head is forged and the bolt or rivet ejected—one on each stroke.

The continuous motion automatic feed machine, due to the mechanical feed, is considerably more productive and it gets the call over the hand feed machine where production requirements are high. It is operated in the same manner as the hand feed, except that mill length bars are used. The entire length of the bar is heated and the operator merely starts it into the feed rolls, subsequent operations being entirely automatic.

Fig. 12 shows a continuous motion automatic feed bolt and rivet forging machine. The hand feed machine is exactly the same, except,

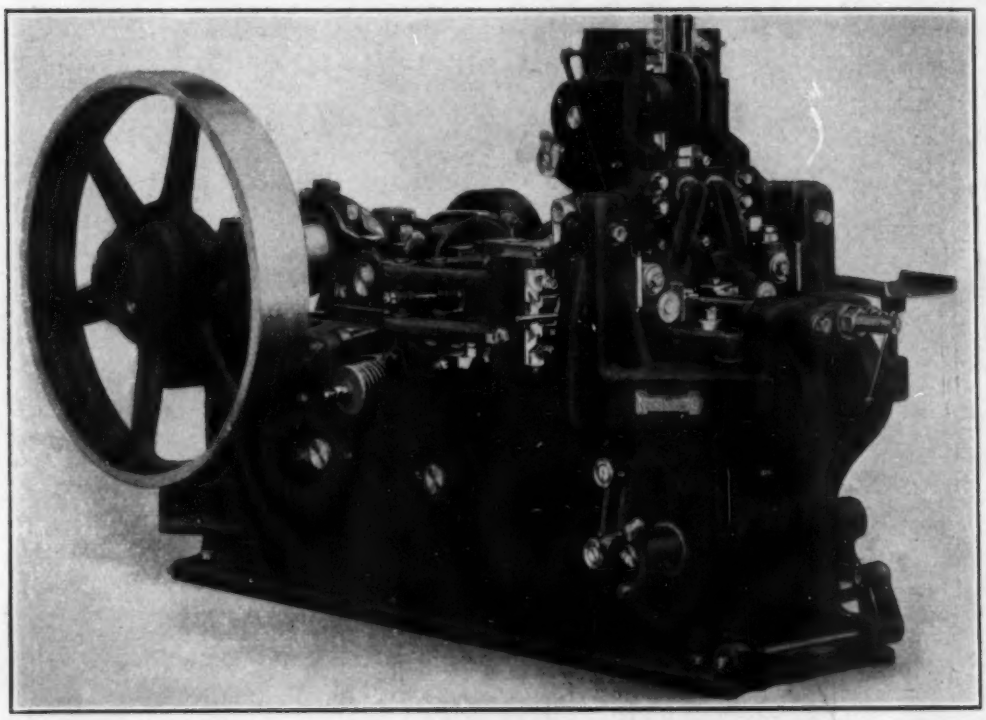

Fig. 14—Photograph of Semi-Automatic Hammer Header.

of course, the automatic feed is left off. Fig. 13 illustrates some of the products of this type machine.

Another type of continuous motion machine is the semi-automatic hammer header shown in Fig. 14. This machine is used for making square, hexagon, and tee-head bolts. It may be operated on long bars like a continuous motion hand feed rivet and bolt header, but the great majority of work produced upon it is made from blanks previously cut to length. A notable difference between this machine and a conventional heading machine is that it has four side hammers which work alternately in opposed pairs. The repeated blows of the

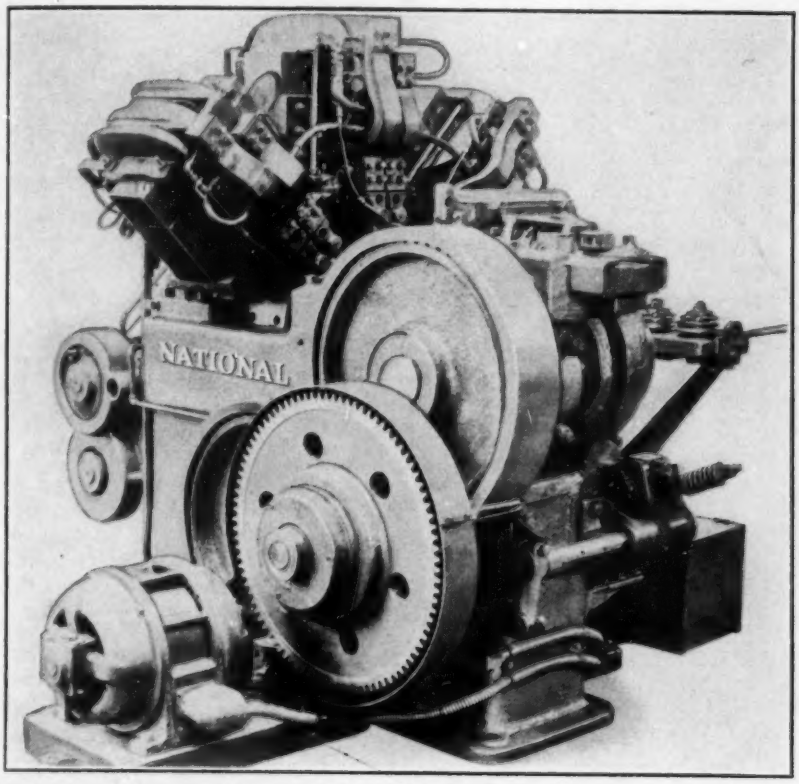

Fig. 15—Photograph of an Electric Header which May be Had in Either Hand or Automatic Feed Types.

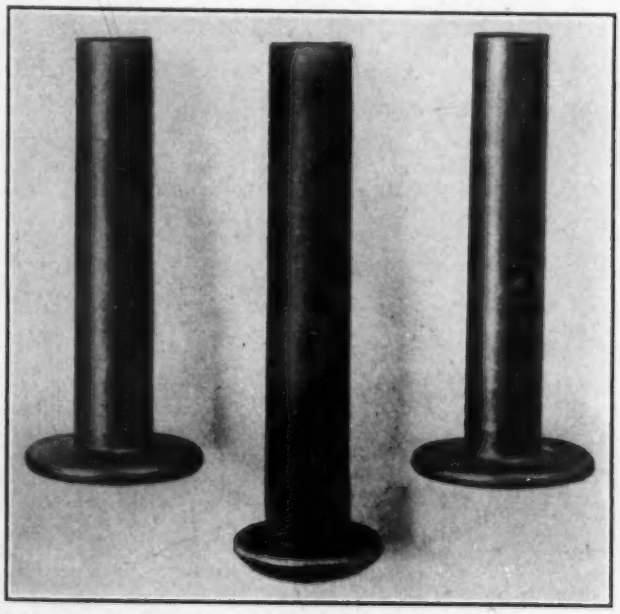

Fig. 16—Samples of Products of the Three-Stage Electric Heater Forging Machine.

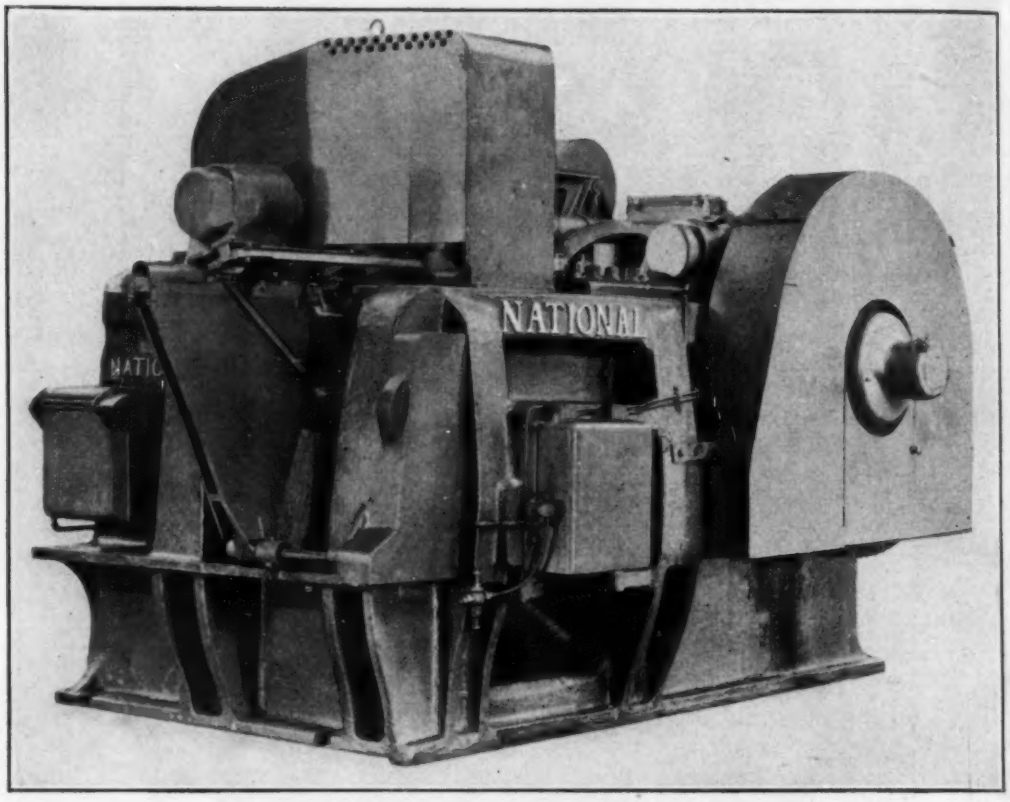

Fig. 17—Photograph of Automatic Feed High Duty Forging Machine.

hammers eliminate all draft and crackle off the scale, thus resulting in a bolt of superior finish. Where some reduction in output is permissible to obtain this better finish, the hammer header is the choice.

Fig. 15 illustrates an electric header which may be had in either hand or automatic feed types. This machine is equipped with a three-stage electric heater. When arranged for automatic feed, the wire is fed directly from a coil, the blank is cut off; and, as it is carried from the cut-off stage to the heading stage in a revolving drum, it is heated by the three-stage electric heater, then headed and ejected. This machine is used for steel parts that can be made in one blow where the requirements call for work having scale-free shanks, the material of which cannot be successfully cold-headed. In this machine the only part of the blank that is heated is that part which is to be upset; consequently, the shank is left in its original state. It is also an ideal machine for handling parts forged from nonferrous stock, because of the accuracy with which the heat can be maintained. Some idea of the possibilities of this type machine may be gained by an examination of the samples shown in Fig. 16. Where production is great enough, certain work is being produced on automatic

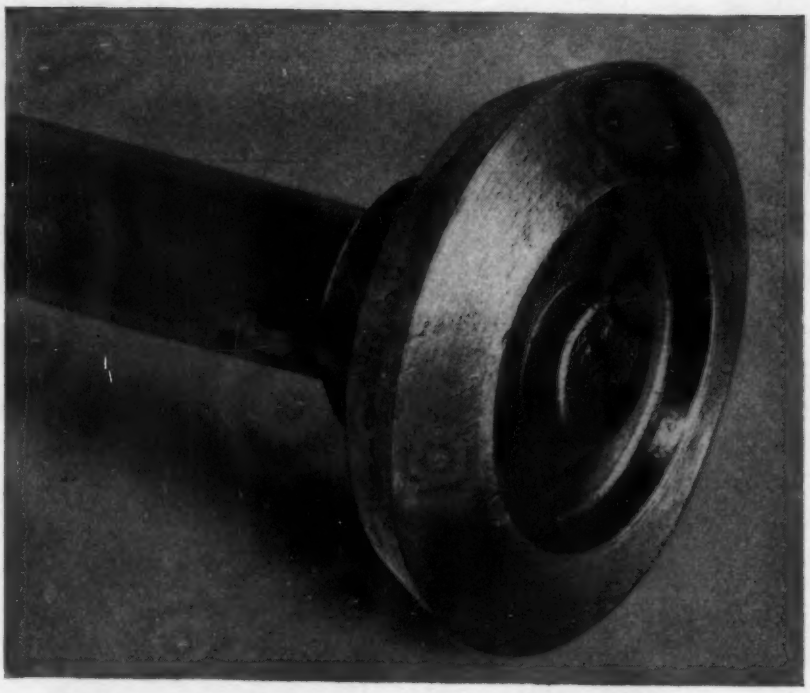

Fig. 18—Photograph of Automobile Axle Shaft Made on Heavy Duty Forging Machine.

feed high duty forging machines. One of these machines is shown in Fig. 17, and the automobile axle shaft pictured in Fig. 18 shows the forging that is being made upon it.

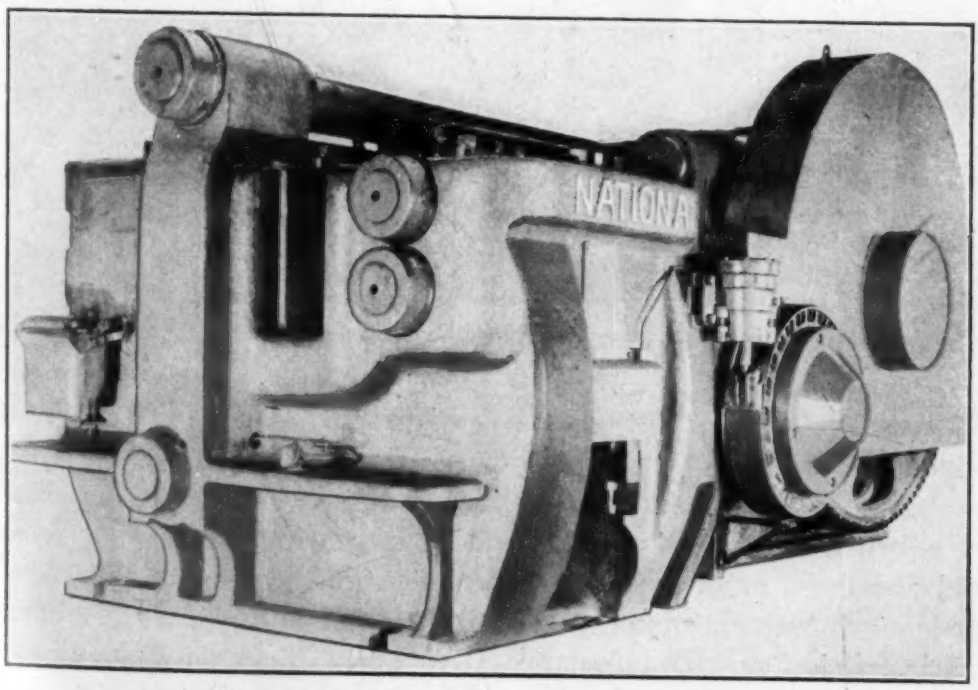

Fig. 19—Photograph of Modern Heavy Duty Forging Machine.

The foregoing equipment is for specialized types of work; and, as previously stated, has been dealt with briefly so as to conserve more space for the modern open die high duty type forging machine. It has a much wider application and is of universal interest.

The forging machine has undergone such pronounced changes

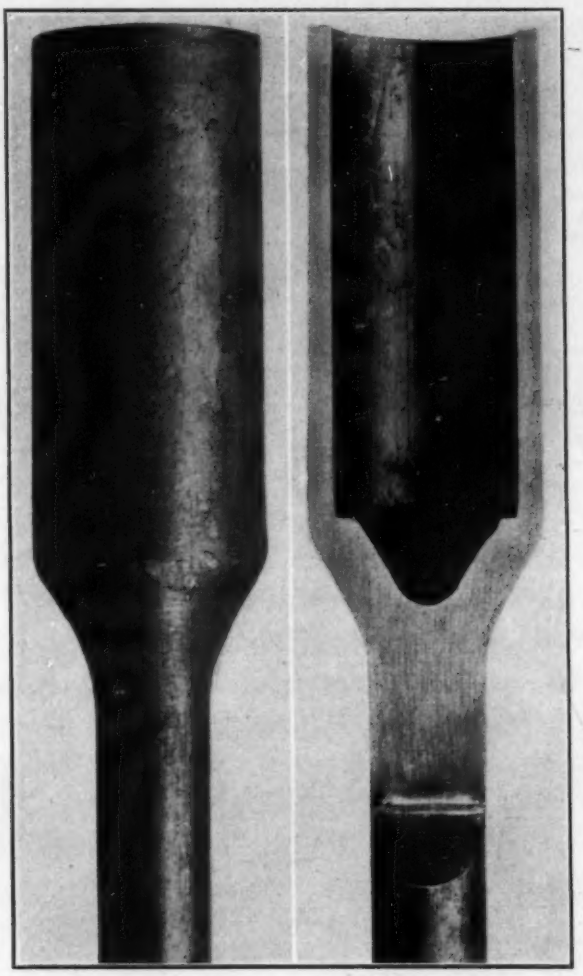

Fig. 20—Photograph of Automobile Drag Link Forging.

in recent years that the old-time forgeman has difficulty in recognizing the modern machine as shown in Fig. 19.

Among its outstanding features are an underslung, stretch-proof, steel bed frame; angular, adjustable main bearings; piloted over-arm ram; under-arm grip slide; suspended slide bearings; double cam gripping movement; interlaced toggles; automatic grip relief; pneumatic friction clutch; automatic safety on heading movement; automatic oiling system; greater power; increased die depth, and more

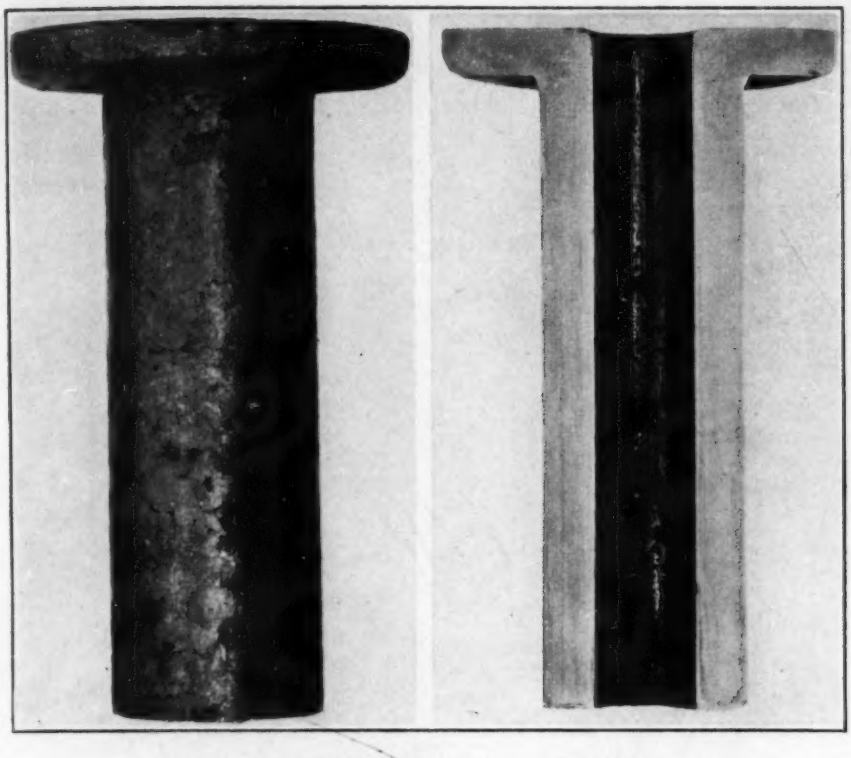

Fig. 21—Photograph of Sleeve Forging.

die opening and gathering capacity. These revolutionary improvements are reflected in accurate output, increased production, lower operating cost, and an almost unlimited widening of the variety of work which can be handled.

The application of the modern forging machine is limited only by the ability and ingenuity of the die designer. Jobs which were labeled "stunts" only a few years ago, are now ordinary, everyday performances. An outstanding example of early pioneering is the automobile drag link forging shown in Fig. 20. Here is a forging which is impossible to make in any except the most modern high duty type machine. In fact, when it was first produced in this modern machine, it was referred to by many experienced, old-time forgemen, as a "trick" forging and a job that certainly could not be economically made on a mass production basis. Another early job in this same category is the sleeve forging shown in Fig. 21. The hole in this job is seven diameters deep. In spite of the dire predictions, however, both jobs were highly successful. In fact, the pierced drag link is still standard on practically all automobiles, well over 10,000,000 of them having been produced.

Continued improvements in both methods and equipment have so

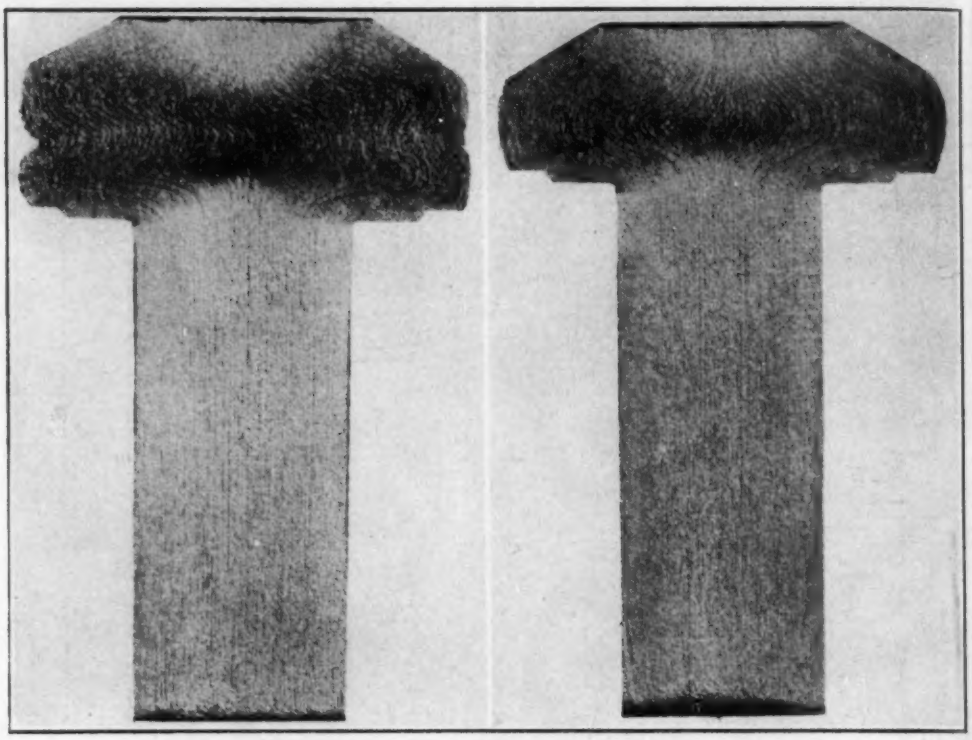

Fig. 22—Left—Photograph of an Upset in Which Improper Die Design has Resulted in Re-Entering Fibers in the Head.
Fig. 23—Right—Photograph of an Upset in Which Proper Die Design has Resulted in Proper Fiber Pattern in the Head.

broadened the field of application that the modern forging machine is daily producing an almost unlimited variety of seemingly impossible work.

Fiber Flow

The subject of fiber flow is receiving the attention of practically all users of forgings. Many forging prints now show and specify fiber flow and grain distribution patterns. Proper fiber flow depends upon correct die design, and no die can be properly approved until a forging has been etched to see if the fiber flow is satisfactory. A thorough understanding of the laws governing forging machine die design, plus their correct application to suit specific requirements, will insure proper fiber flow and guard against forging weaknesses, particularly re-entering fibers. Briefly, these laws govern the amount of stock which may be gathered or upset in various ways in each forging operation without buckling the flow lines and causing defects. Fig. 22 shows an upset in which improper die design has resulted in re-entering fibers in the head. No such defects occur if the die is properly designed, as is shown in Fig. 23.

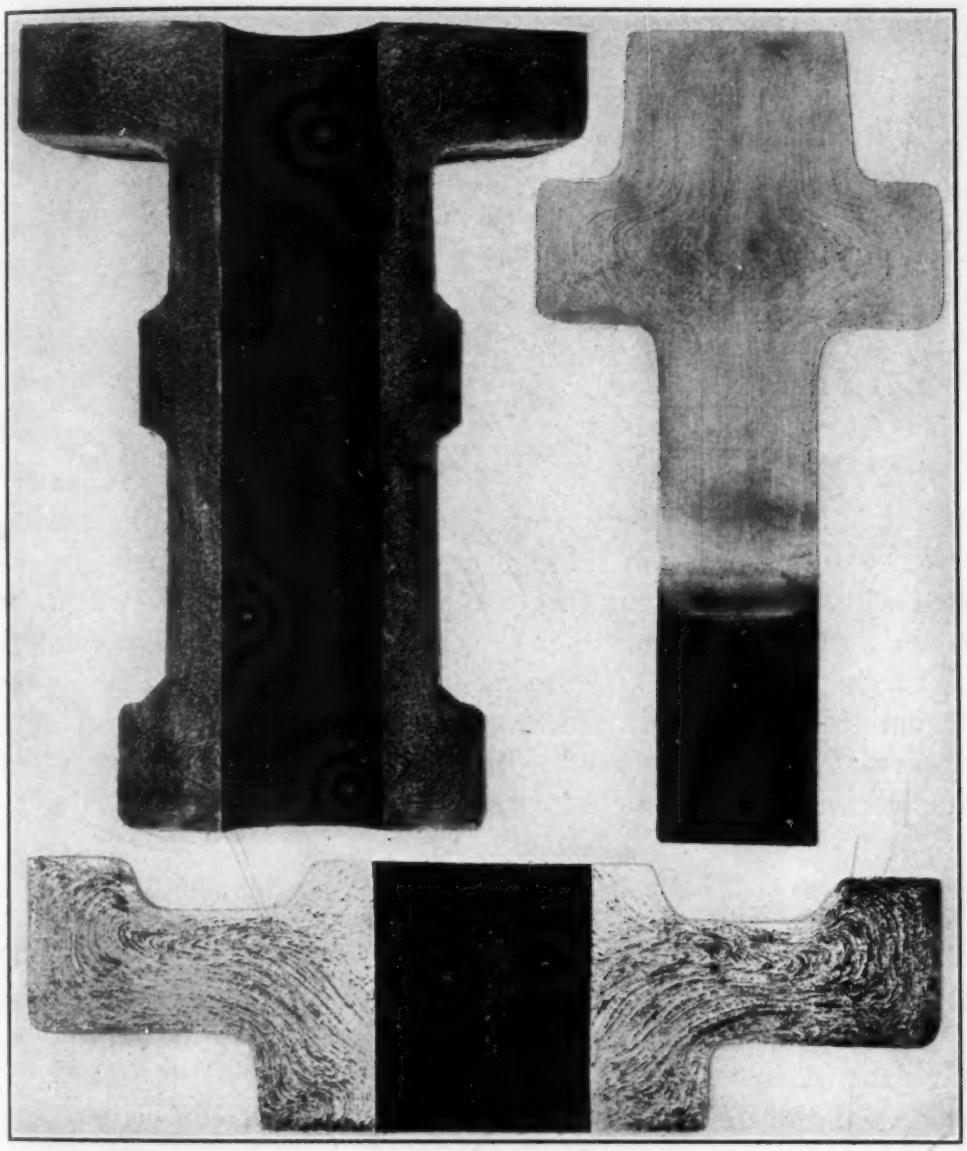

Fig. 24—Group of Etched Forgings Which Have Been Selected to Cover Jobs of Widely Varying Types. These Have Been Sectionalized and Deep Etched to Show the Fiber Flow.

Fig. 24 shows a group of etched forgings which have been selected to cover jobs of widely varying types. While the dies required for these various forgings were entirely different, it will be noted that close adherence to correct die design rules has resulted in sound forgings with proper fiber flow lines. It may be readily seen, consequently, that fiber flow is something over which the machine manufacturer has little control.

Grain size for machinability and warpage is also a factor that can be controlled only by the user. Investigations along this line

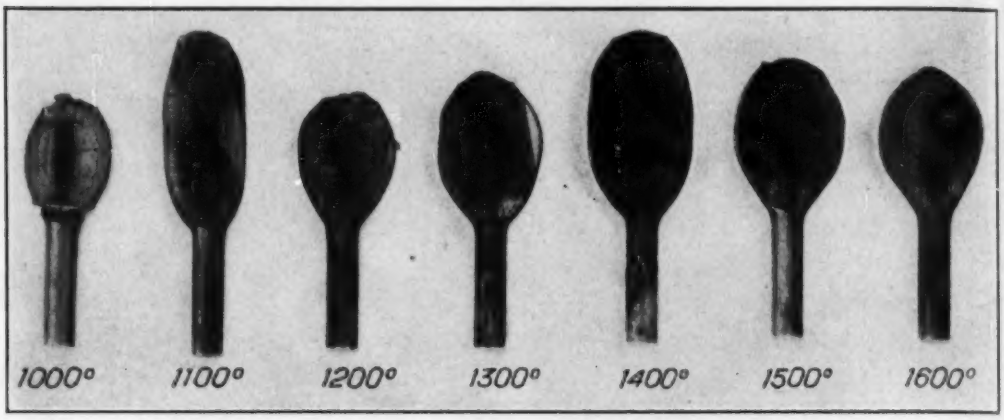

Fig. 25—Illustrates a Temperature Test on a Forging Brass.

seem to indicate that a dense forging machines easier and there is less tendency for warpage in subsequent heat treatment than is encountered with a more porous forging. There is also some evidence to show that density depends upon the character of the steel from which the forging is produced, the temperature at which it is forged, and the design of the dies. However, much research work is still in progress; and until all phases of this broad subject have been covered, complete and definite conclusions cannot be established.

Temperature

Temperature is another factor that is entirely under the control of the user. In general, the temperature of the stock is determined by the shape and character of the forging, the ease or difficulty in filling the die recesses, the kind of steel from which the forging is made, and the finish desired. If the forging has off-center lugs, undercuts, crevices, or other characteristics which make it difficult for the stock to flow into the die recesses, or a large number of operations are required to complete the forging, the temperature of the stock must be comparatively high. On the other hand, a simple forging, such as a bolt or rivet, can be made at a low temperature. The temperature at which most steels begin to scale appreciably is at about 1500 degrees Fahr.; consequently, parts forged at a temperature below this point are free from excessive scale. It is also well known that proper furnace design, control of atmosphere, etc., assist in the reduction of scale.

Nonferrous stock must, of course, be forged at much lower temperatures than steel, and must be held within a very narrow range.

Fig. 25 illustrates a temperature test on a forging brass having an analysis of copper 60 per cent, zinc 38 to 40 per cent, and lead 2 per cent maximum—the lead improving the machining qualities; and, at 2 per cent maximum, having no detrimental effect upon forgeability.

As will be noted, specimen No. 1, which was forged at a temperature of 1000 degrees Fahr., is checked to such an extent as to clearly indicate that the stock will not properly forge at this low temperature. Specimens No. 2 and No. 3, forged at temperatures of 1100 and 1200 degrees Fahr. respectively, are also badly checked, making it evident that 1200 degrees Fahr. is still not a sufficiently high heat to cause the material to flow properly. Specimen No. 4, which was forged at a temperature of 1300 degrees Fahr., shows a fairly good flow of the metal, but even at this temperature there is some tendency for checks and cracks to develop at the juncture of the flattened portion and the stem. Specimens No. 5, No. 6, and No. 7 were forged at 1400, 1500, and 1600 degrees Fahr. respectively; and, as may be seen, the stock flows quite well at any of these temperatures. The bar melted at 1650 degrees Fahr.; consequently, 1600 degrees Fahr. is the absolute maximum temperature at which it is safe to heat this material. The practical application of this test shows that perfect results are obtained if the temperature range is held within 1425 and 1475 degrees Fahr.

Aluminum is forged at considerably lower temperatures, and also must be held within close range, depending upon the alloy. As an example, certain standard aluminum alloys forge best at temperatures somewhere between 800 and 950 degrees Fahr.

In the final analysis, no hard and fast rules can be established for forging temperatures, either on steel or nonferrous stock. They can be arrived at only after all the contributing factors of each specific job are known and considered.

DISCUSSION

Written Discussion: By L. W. Davis, Aluminum Co. of America, Cleveland.

Mr. Friedman is to be congratulated for his clear description of recent developments and improvements in both forging presses and upsetters. It is only by the use of these improved machines that many of the finest specimens of the forging art have been produced. This is especially true in the nonferrous field where the practical absence of scale and the lower forging temperatures make a press particularly attractive.

In this field it seems that one of the principal problems is that of die life.

Much can be accomplished to improve the life of pressing dies by refinements in the design of the impressions in the die to decrease resistance to the free flow of the hot metal. We have in mind one particular hot pressing job in which the die life was increased from an average of 400 to 500 pieces to an average of about 5000 pieces by some very simple refinements in die design which did not affect the dimensions of the piece supplied to the customer. However, if a part of this same type were produced in a drop forging hammer, the life of an impression would be at least 10,000 to 12,000 pieces and resinking in the same die block would be possible.

It may be that further refinements in die design would increase the life of the pressing dies but it seems more probable that other die steels or other heat treatments of the same die steel are the lines of attack most likely to produce results. Of course, the users of forging presses are working on this problem continuously, but those of us who are in the nonferrous field are not experts on special steels. A co-operative attack of this problem by the builders of hot pressing equipment and the makers of hot working die steels should produce some very desirable results. Marked improvement in die life would certainly result in an immediate widening of the application of the forging press in the nonferrous forging industry.

Written Discussion: By A. V. Wetzel, metallurgist, Titan Metal Manufacturing Co., Bellefonte, Pa.

This paper arouses some very interesting points with regard to die steels for hot forging work, and mention was made by the author of the fact that he has found it to be the best practice to heat treat or quench the dies from the high temperature of the permissible hardening range for a two-fold purpose to produce a coarse grain in the die, and to also raise the temperature at which tempering is performed.

The writer should like to ask Mr. Friedman if he has found any relation between austenitic grain size and the resistance of the die to heat checking, particularly in the tungsten die steels of the semi-high speed type. We have done some work along those lines, but could learn nothing very definite as far as die life is concerned in hot forging of brass at temperatures of 1350—1400 degrees Fahr. The tendency seems to be toward a finer grain for best die life, but this statement may not be considered as conclusive.

Oral Discussion

I. E. WAECHTER:[1] What, in Mr. Friedman's experience has been the most satisfactory hardness for forging dies, as well as break-down and coining dies to obtain most satisfactory die life?

J. H. FRIEDMAN: On the average, the hardness of die should be as great as can be used on the type of die that is under consideration. In other words, a die that has projected areas of quite some extent is subjected to bursting pressures, consequently must be very much softer, of course, than a die that is the other extreme, perfectly flat. On the average, for experimental purposes and development work, a concern will start using a die that is barely machinable. Such a die will be 55-57 scleroscope hard and will never break. From there,

[1]Metallurgist, Weatherhead Company, Cleveland.

observations are made as to how much harder the die can be made, and the hardness is built up to a point where fracture occurs infrequently. There is a more important thing than hardness in getting die life, and that is density of the material, and surface hardness. A good hot-work steel might well be only 55 or 60 hard and still appear to have a hardness very much in excess of that. After working on hot work, it will develop a very much harder surface after the die is properly broken in. It gets a good, shiny black slick surface appearance, and will acquire quite a great hardness on the surface, but it will be very tough within. That is about what happens on a good hot-work alloy that is heat treated up to 55 hard to begin with, and that, on the average job, is where I would suggest as the place to start.

I. E. WAECHTER: What temperature does the surface of the die attain while a forging is actually being fabricated?

J. H. FRIEDMAN: That would be very difficult to state except as an estimate, and the estimate has to be made as the result of a notation of the effect of the hot work on the various steels, knowing what the steel will stand.

In the first place, it would be necessary to know the temperature at which the forging is being made. We will assume that the forging is being made at 2000 degrees Fahr. That forging, under a very heavy pressure, will develop a temperature on the die based on how long it is in contact with the die, and the pressure of the die on the forging. Also, we have to know the size of the forging, and the thinness of the forging. In other words, we have to estimate the pressure of the forging as well as the temperature, and the length of time in contact with the die. It is certainly not at all unusual on these modern machines to see forging dies actually running in a red hot condition. The forging dies on certain work are heated to a deep cherry red point easily seen in daylight.

I. E. WAECHTER: You would say that a steel possessing good hardness in addition to good toughness at those temperatures would be the most applicable one to good die life?

J. H. FRIEDMAN: The most destructive things on a die are heavy pressure and abrasion, and particularly heat,—the heat interferes with the hardness. If a die is made too hard heat checking occurs when in forging the surface is caused to expand quickly while the rest of the die does not expand. In other words, it is a localized heating, and the surface of the steel has to be in condition to expand and contract repeatedly without heat checking. When you get a die that does not heat check and has a good resistance to abrasion — which, as a rule, is brought about through a little molybdenum in the steel, and has a satisfactory red hardness — you usually have the basis of a good die steel for hot work.

H. G. KESHIAN:[2] I wish to ask Mr. Friedman a question regarding the best temperature for hardening hot forging dies. As he undoubtedly knows, some people prefer high and others low temperatures for hardening the similar types of steel. These temperatures may vary say from 1900 to 2200 degrees Fahr. I would like to know if Mr. Friedman has had any experience in this respect, and if so, has he noticed any material difference in the performance of

[2]Metallurgist, Chase Brass and Copper Co., Waterbury, Conn.

dies made from the same steel and hardened from high and low temperatures.

J. H. FRIEDMAN: There will always be a difference of opinion on the correct hardening temperature for dies, but it has been our experience that on the average, the dies should be carried through the full heat even to a point of over-heating for grain growth, rather than the under-heating treatment. Also, on the draw we much prefer to heat through the secondary hardness range on the tungsten steels. If you do not get through that secondary hardness range, the die will go through it in the process of doing the work. In other words, you are applying heat at drawing temperatures, which will cause that die, if it has not gone through the secondary hardness point, to become harder in use, and because of this secondary hardness it will heat check on the surface.

There is a secondary hardness range at which you will increase hardness materially — five to seven points on the scleroscope, which would make it much too hard.

I have heard of some good results on certain types of steel from heat treatment of the tungsten steels in 1900 degrees Fahr. range, but I do not believe that on the average it would be a good treatment to recommend. Certainly it is worth playing with on high production work, but it could not be recommended in general. In other words, our experience has been that over-heating through the critical is better than any under-heating treatment that we have found.

Frequently a lower hardening range is used to get away from a draw. There seems to be the ability to produce a tougher die at the lower hardening point, which will get by better without a draw than is possible by going through the upper temperature in the hardening.

In other words, you can get a tougher die, and you can get by better without drawing the die if you use the lower temperature in hardening; but, in general, I am afraid that the lower temperature would be considered a freak treatment.

E. F. GIBIAN:[3] It has been our experience in extrusion processes that dies which had been hardened to a Rockwell C hardness of 54 maintained their hardness even after many thousand pieces, indicating that the temperature of the die never went above 1000 or possibly 1100 degrees Fahr. In order to get enough strength in some very difficult jobs, we had to raise the hardness, and to overcome brittleness, we used a tungsten steel die of a very small diameter, and shrunk it into tougher steel rings.

Another possibility of overcoming the failure of dies, is to apply a short air blast between each forging operation.

I wonder if Mr. Friedman has any experience with cooling the dies by oil spray or air spray where severe service of the die is required.

J. H. FRIEDMAN: A good many people, particularly on that most interesting job of forging automobile drag links, use an oil spray in which they atomize the oil, on the die, through what you might term a spray paint gun.

Other people just use an oil drip to cut down the abrasion; but, as a cooling medium, I know of no successful application on die steels other than pumping a liberal amount of soluble oil on the dies and tools or internal water

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[3]Thompson Products Co., Inc., Cleveland.

cooling of dies by jacketing, — cutting grooves around the inserts, as you speak of. There are some very successful applications on insert dies, where grooves are formed around the insert on the holder, through which circulates a quantity of water. That last is the effective way to cool the die.

Any cooling on the surface other than soluble oil, I believe, would be prohibitive. Air would be entirely too expensive, — at least I believe it would be considered so by the average forge man. Water would be destructive, and oil would be dangerous because of its inflammable condition in quantities. Perhaps a better scheme of cooling dies, and one we would recommend, would be water jacketing of these inserts in the holder. This inserting is a very interesting subject. It is being very broadly adopted by both the cold and the hot industry.

Inserting is a matter of putting a plate on a die, or putting a bushing in a solid die, the dimensions of the holder being such that it overpowers the bushing. In other words, the holder must cause the bushing to contract when it is pressed or shrunk into the holder. I like the pressing operation better than shrinking. Care must be taken not to overload the bushing to a point where its compressive strength has been exceeded, otherwise the insert will be fractured. The holder must be large enough (larger than even your mind's eye pictures it in diameter) so that, under the application of the forging pressure, the holder never permits the insert to go beyond its original normal diameter because the insert will not stand anything in tension. On the average you will find it good practice to use a holder three to four times the outside diameter of the insert. If you have an insert with a wall thickness that balances it with the holder, and a holder large enough to cause it to contract from its original diameter, the holder is then so strong that the forging pressure will not open it up beyond its normal diameter in the hole.

It is a very interesting subject, and it is being broadly tried and adopted in many of the large production plants.

There is also another point that is interesting. A good many people are making dies in pieces. You will find that you can now secure round dies for heavy pressure work made from tungsten carbide material.

General Electric Company is doing some very interesting work in furnishing split dies made from tungsten carbide for such things as tube drawing. You will agree that is fine workmanship. It is so well done—put into a holder under such a pressure—that the joint is absolutely not noticeable. If you can notice a joint you are "sunk". The metal will flow into it and ruin the die. However, it is being very generally done with excellent success. Break your die to begin with, where it is going to break anyway, do good machine work, and you will be all right.

Author's Reply

Answering Mr. Davis' written discussion without knowing the design and shape of the die job in question and the rate of heat transfer to the die, it becomes impossible to make a definite answer. However, I might state that various jobs have been brought to our attention on nonferrous metal where the die life has been increased from 5000 forgings, as previously made, to 20,000, and in some cases even 50,000, forgings when made on a Maxipres. Obviously, it

is impossible to comment on specific die cases without knowing all of the details, but the fact must be kept in mind that when a job is transferred from the hammer to the press it becomes a totally different kind of a forging. It no longer has draft, and the closer tolerances permissible on the press result in a forging that is much nearer the final size and shape of the finished part. Under such conditions, even though die life may be longer on the hammer, the ultimate cost is much less if the forging is made on the press, because of the saving in material, and the reduction in heating and machining costs.

In answer to Mr. Wetzel's question, the relation between austenitic grain size and the resistance of the die to heat checking is a very debatable subject. Very little is known about the mechanism of heat checking. Tungsten die steels of the semi-high speed type, such as used for hot work, approach the "precipitation type" of hardness. Steels with higher tungsten or higher carbon fall into the "martensitic type." There is a definite difference in the reaction of these two types to heat checking. As a general statement, it may be said that particularly where shock is to be resisted, the high temperature treatment producing a coarse austenitic grain size seems to resist heat checking, as well as the result of heat checking, which is the rapid fatigue fracture starting from the heat check. I know of large users of tungsten hot work forging die steels who order steel with an ingot structure. The ingots are cast very close to the desired size and very little work, as far as forging, is done. This bears out the coarse grain theory and has established in my mind an opinion that the austenitic grain size is definitely related to satisfactory performance. No definite data on the subject, other than my opinion based on experience, can be expressed.

EXTRUSION OF METALS

By D. K. Crampton

Abstract

The term "extrusion" as applied to metals and alloys comprises essentially the application to a relatively massive billet or blank sufficient pressure to cause the metal to flow through a restricted orifice, thereby forming a greatly elongated section of uniform but relatively less massive volume. The extrusion may be either hot or cold. As the terms are here used, hot extrusion indicates that the metal is above its re-crystallization point and thus continuously re-crystallizes and is not work-hardened during the process. Cold extrusion indicates that the metal never reaches the temperature of incipient recrystallization. It may, therefore, be considerably work hardened by the process.

Extrusion of lead and some lead alloys was introduced at about the beginning of the nineteenth century, and for many years only low melting point soft alloys were so treated commercially. It was not until near the close of the nineteenth century that a copper-base alloy was successfully extruded at a relatively high temperature. During the last twenty years, especially, great strides have been made in the equipment, processes, handling and treatment of alloys during and subsequent to the extrusion process. As a result, many alloys and shapes are now extruded commercially which heretofore were considered impossible or impracticable.

Both mechanical and hydraulic extrusion presses are used but the tendency today is almost exclusively toward the hydraulic type. These are more flexible, compact, easier to control, and less subject to damage from improper operation. The presses most used at present vary in capacity from about 200 tons to 4000 tons.

Hot Extrusion of Lead and Lead-Base Alloys

THE ordinary extrusion of low melting point alloys such as lead, tin, zinc, lead-tin alloys, pewter or britannia metal, etc., is usually carried out either at room temperature or at least fair-

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ly low temperatures. The principal applications of lead and lead alloys are for cable sheathing, pipe, solders, hose encasing and mouldings. Of these the production of cable sheathing is distinctly the most important from the standpoint of volume. There are two general types of cable as regards field of use that are lead-sheathed; first, telephone cables, and second, power cables. Most telephone cables are suspended and, therefore, subject to wind and ice loads and to considerable temperature variation. Power cables are mostly carried through ducts and, therefore, not subjected to such severe service conditions.

Up until about the present time telephone cable production has required enormous quantities of lead for sheathing, but the amounts so used have fallen off considerably, due partly to general poor business conditions and partly to improvements in telephonic communication, whereby many more messages can be sent simultaneously over a given pair of wires than heretofore. This latter applies only to main trunk cables between large cities.

In general, lead sheathing for telephone cables has been produced from some type of hardened lead, mostly antimonial lead. This antimonial lead usually contains approximately 1 per cent antimony, which renders it susceptible to precipitation hardening, although advantage is not taken of the full hardening effect on account of a consequent tendency toward brittleness. Frequently this antimonial lead is blanketed as it is extruded so that a moderately slow cooling results. This gives an intermediate hardness and a relatively low aging effect.

Many other types of lead have been or are being investigated, some of which might be found to give a better combination of properties for cable sheathing than the antimonial lead. Typical of these newer leads are those containing:—

A. Approximately 0.06 to 0. 08 per cent tellurium
B. Approximately 0.25 per cent cadmium with either 0.5 per cent antimony or 1.5 per cent tin.
C. Approximately 0.06 per cent copper.
D. Approximately 0.04 per cent calcium.

Lead sheathing for low and intermediate voltage power cables when carried in ducts for distribution service is normally produced from soft lead as the conditions of use do not necessitate the resistance to fatigue and creep which is so important in telephone cables. On the other hand, there is an increasing use of

high voltage oil-filled power cable where the requirements are a little more drastic and where a sheathing lead of higher strength and creep resistance is required. At present, lead containing approximately 1 to 3 per cent tin is commonly used, but here also other alloys are being carefully examined. This alloy is commonly quenched as extruded. Power cables for aerial service, however, usually have sheaths containing either about 0.75 per cent antimony or 2 per cent tin. Leads containing 0.06 per cent copper more or less show some tendency to replace the tin and antimony bearing leads. The indications are at present that the creep resistance of the copper-bearing material is better and the cost materially less.

Although it has been realized that fatigue and creep resistance are very important in many cable sheath applications, adequate and reliable creep data on the various types of lead have to date been lacking. As this is being written a paper is scheduled to be given before the June meeting of the A.S.T.M. by A. J. Phillips which will contain many new data on various leads. For those who are particularly interested in the comparative physical properties of the various leads, reference to this paper is recommended. For present purposes, it would seem sufficient to quote short time tensile tests which will give an approximate indication of the hardening effect of some of the various alloy leads referred to. In quoting these values particular attention is called to the fact that there is considerable creep in all of these materials at stresses considerably below those quoted so that in no case should these data be used for design purposes. The following data are the result of ordinary tensile tests carried out at 25 degrees Cent. at a speed of 0.05 inches per inch per minute on strips 0.10 inch thick x 1.0 inch wide and which had been prepared by rolling by a series of 10 per cent passes cold, care being taken not to permit an appreciable rise of temperature during rolling.

Type of Lead	Approximate Short Time Tensile Strength Pounds Per Square Inch
Soft Commercial Lead	1800
Chemical or Copper-Bearing Lead	2300
Tin-Bearing Lead	3000
Antimonial Lead	3200

The pressures required to extrude will vary considerably with the alloy as well as with the temperature, size of section, and num-

ber of sections per die. Soft lead usually will require unit pressures from about 40,000 to about 60,000 pounds per square inch. For any given section the copper-bearing leads will require roughly 10 to 20 per cent more pressure, and the antimonial and tin-bearing alloys possibly 50 per cent more.

The speeds permissible also vary greatly. With soft and copper-bearing leads and large sections, this is limited only by the capacity of the pumps. In very small sections it may in some instances run as high as 200 feet per minute. However, with the other alloys

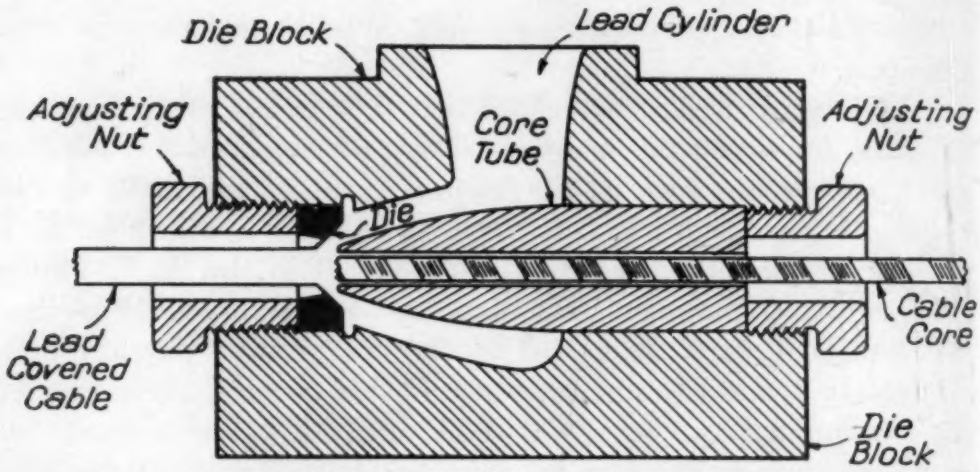

Fig. 1—Section Through Die-Block Assembly Used in Making Lead Cable.

mentioned, speeds of from 20 to not over 80 feet per minute are necessary to prevent checking. The physical properties of the sheath vary somewhat with the speed, better values resulting from the lower speeds.

The equipment for extruding cable sheath includes a melting kettle with connection to the extrusion press. The usual press is hydraulic and so arranged as to take intermittent charges of molten lead from the kettle. This press is vertical, and the cylinder in a commercial machine might hold approximately 1800 pounds of lead. At the lower end of the cylinder, there is a die block forming an annular path for the flow of lead around a core through which is fed the cable to be sheathed. A section through a die block assembly is shown in Fig. 1. The molten lead is charged into the press at about 750 degrees Fahr. and it normally requires approximately 7 to 8 minutes for solidification and reduction to a suitable temperature for extrusion (about 425 degrees Fahr.) at which time the pressure is applied. The now solid lead is forced through an annular

orifice between the core block and the die and on to the cable, the product normally being coiled on a large drum as it comes from the press. The presses commonly used for this work have a capacity of from about 600 to 2500 tons and can produce up to 5500 pounds per hour or slightly better.

One of the difficulties encountered in the finished cable sheathing is a seam due to entrapping oxides or other foreign materials between successive charges of lead. There is, of course, at all times a layer of dross on the surface of the molten lead and even with the utmost care, some of this may be entrained, tending to prevent perfect coalescence of the divided lead stream as it flows around the core block or at the interface between two successive charges of lead fed into the cylinder. When an appreciable quantity of such dross is occluded a seam or a weak spot results, which is likely to fail in service. Much study is now being given to various means for more positive prevention of dross inclusions, as for instance, purging of the feed line with carbon dioxide or other nonoxidizing gases, so-called continuous feeding from the kettle to the press and in one commercial installation the central melting and continuous circulation of molten lead to a battery of presses. With such improvements and greater care and supervision, this difficulty is being largely overcome in present day practice.

A very recent development in cable sheath extrusion has been brought out at the Pirelli-General Cable Works in England. This is a radical departure from the older extrusion processes in that the molten lead is fed to a chamber containing a revolving steel screw by means of which the lead is forced ahead. It is simultaneously cooled and solidified. At the end of the screw the lead is forced through a die to form a sheath on the cable. This part of the operation is exactly the same as in the ordinary press. These screw presses are made in three sizes; the largest taking up to 5-inch diameter cable. This large machine has a capacity up to 4 tons of metal per hour, and the smallest machine about 0.8 tons per hour. The output is somewhat higher than that obtainable from the ordinary hydraulic press as it is a continuous operation, not intermittent. It is claimed that this type press prevents the inclusion of oxide and foreign matter which might otherwise cause seams.

(b) Hose Encasing. In the manufacture of rubber garden hose and of rubber covered wire it is common practice to vulcanize the rubber while it is encased in a sheath or mold of lead. This lead

mold is extruded over the hose in quite similar manner to that used in the production of cable sheath. In the present instance, precast lead billets are fed into the press instead of molten lead. This is possible because no particular precaution against forming of seams or the inclusion of dross need be taken. By eliminating the necessity of waiting for the molten metal to freeze, approximately double the production is obtained.

While cold billets may be fed, they are preferably heated to about 260 degrees Fahr. The container itself ordinarily is heated to a temperature of approximately 350 degrees Fahr., and the actual temperature of the lead, as extruded, approximates the latter figure. Pure soft lead is ordinarily used, and this is extruded as fast as the equipment permits, speeds up to 100 feet per minute being attained. After the lead encased hose is taken from the press, it is vulcanized by hot water under pressure which varies from 150 to 250 pounds pressure. The lead is subsequently stripped from the hose and is remelted for further use in the same operation. Under ordinary conditions billets weighing approximately eleven to twelve hundred pounds are used, and the lengths obtained depend upon the size of the hose encased. The lengths might run from 250 to 1000 feet in normal production.

(c) Lead Pipe. A considerable amount of lead pipe and somewhat lesser quantities of tin pipe are extruded. In former times lead pipe was used quite extensively for plumbing and similar services but the amount so used is becoming less and less as time goes on. Particularly with the advent of the several processes for making copper and brass pipe, and copper water tube, the use of lead has fallen off. Although lead is good in most waters from the standpoint of corrosion, it is structurally weak, and will not withstand high pressures. Common or soft lead is the variety used for most pipe but there is some demand for an antimonial lead of 3-6 per cent antimony for use in acid plants.

Vertical presses are used for extrusion of lead pipe. Usually, but not always, the lead is poured in and solidified before extrusion. The die is held in the hollow ram and the extruded pipe comes out at the top of the press and is coiled as produced. The presses commercially used normally run from about 400 to 1000 tons capacity and the individual charges vary from 200 to 500 pounds depending upon the size of the pipe made. Soft lead can be extruded quite fast, the speed being determined largely by the capacity of the pumps.

Speeds of 50 to 60 feet per minute are not uncommon in practice.

(d) Candle Wick Filled Lead Gaskets. A considerable amount of candle wick filled lead gaskets of variable cross section are used for packing joints in concrete water pipes. The metal used is soft lead which is extruded in the form of irregularly shaped hollow sections. A cotton wick is fed in so that it entirely fills the inside of the section. The process is quite similar to that used for cable sheathing or hose encasing. Here cotton wicking replaces the cable or the rubber hose as the inner member.

Hot Extrusion of Tin-Base Alloys

Various solder alloys are extruded in the form of wire. These are mostly tin-lead alloys containing from about 40 to 60 per cent tin. Many other tin-base alloys are produced in lesser amounts. A large part of this production is in the form of solid wire, but a considerable amount also is resin or other flux-cored solder. In the case of the solid wire, solid billets are extruded in an ordinary rod extrusion press, but in the case of flux-cored solders, resort is made to a procedure roughly similar to that employed for hose encasing or cable sheathing. The presses used for solders may be either vertical or horizontal but are small, ranging from about 200 tons to 500 tons and using billets about 2½ to 4 inches in diameter and 8 to 16 inches long. Resin core solders are extruded in a single strand, but solid wires in multiple strands, sometimes as many as 12 per die.

Most solder alloys must be extruded at fairly low speeds to prevent checking of the surface, about thirty feet per minute being usual. The pressures and temperatures run about the same as in the cable sheath extrusion.

A considerable quantity of various ornamental shapes are extruded from pewter, or as it is sometimes called, britannia or white metal. The alloy is of variable composition, of tin-base, and normally contains anywhere from 5 to 15 per cent antimony, from 0 to 3 per cent copper and 0 to 15 per cent lead. These extruded shapes are used for mouldings or "mounts" as they are called in the trade, on various pewter vases, pitchers, etc. They are attached to the plates by the use of low melting point solders.

An ordinary horizontal rod press of about 250 tons capacity is used, the billets being about 2½ inches in diameter by 8 inches long. The alloy is extruded at a temperature of 140 to 150 degrees

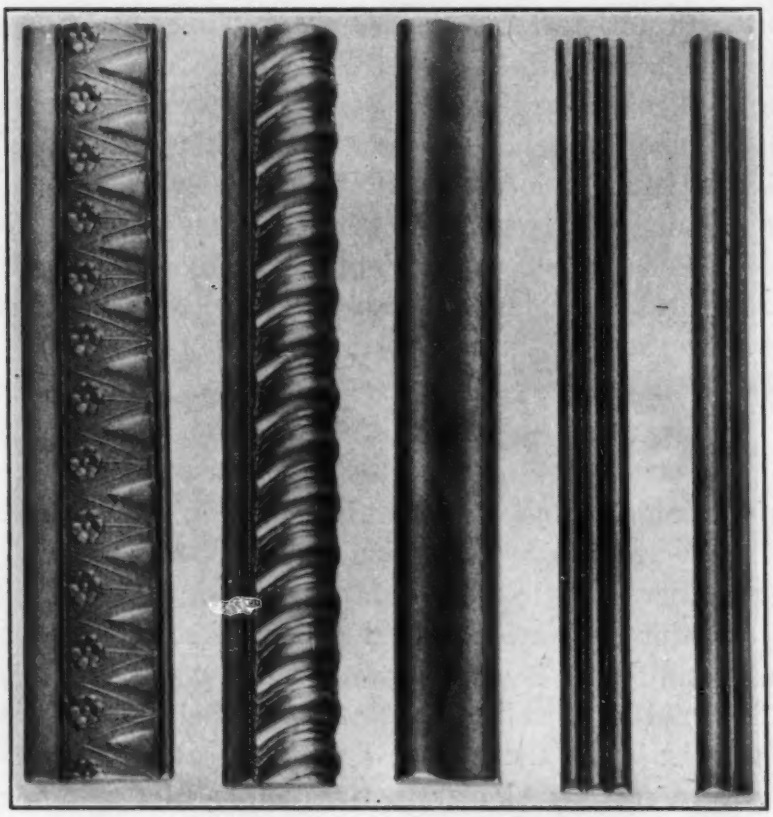

Fig. 2—Extruded "Mounts" in Both the Extruded Condition and After Passing Through a Pattern Roll.

Fahr., but very low speed must be used to prevent checking of the edges. In some cases this might be as low as 10 to 20 feet per minute. The pressures used will vary from 40,000 to 100,000 pounds per square inch.

Where the cross section of these "mounts" is uniform along the length they are used in the condition as extruded. Sometimes a beaded or embossed section is desired in which event the uniform shape as extruded is subsequently fed through a "pattern" roll. Both types of "mounts" are illustrated in Fig. 2.

Hot Extrusion of Copper and Copper Alloys

An extremely large tonnage of copper-base alloys is extruded annually for a wide variety of ultimate uses. The equipment used is different in many respects from that employed for the low melting point alloys although the principles involved are much the same. Both mechanical and hydraulic presses have been used but the mechanical ones are much in the minority and the present tendency is

entirely toward the hydraulic. The relatively few mechanical presses in use today are employed for small tubes, not for rod, and are of relatively small capacity. A typical horizontal hydraulic press in operation is shown in Fig. 3.

Rods are always extruded from horizontal hydraulic presses, but for tubes both horizontal and vertical ones are used. For tube extrusion the smaller presses of capacities up to perhaps 1100 tons are prevailingly vertical and the larger ones invariably horizontal. Tube presses up to 4500 tons are in use and rod presses up to about 3500 tons.

European and American practices differ with respect to the use of accumulators. They are almost always used both for rod and tube in European plants. In America the use of an accumulator is rather

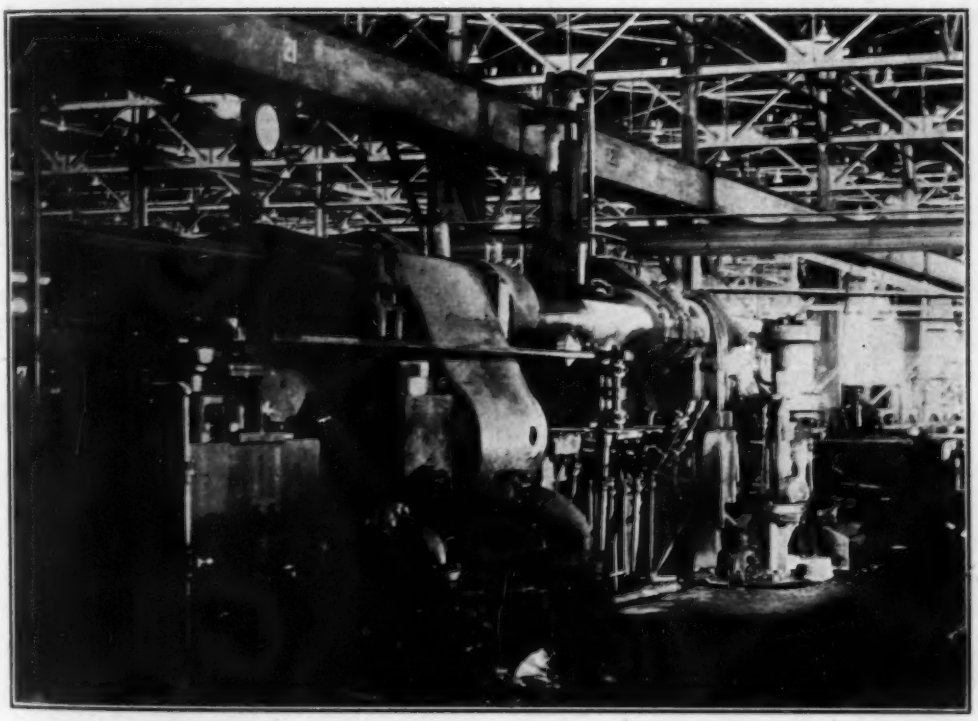

Fig. 3—Photograph of a Typical Horizontal Hydraulic Extrusion Press Used in Extruding Copper and Copper Alloys.

rare in connection with rod presses but quite common in connection with tube presses. This is because most rod alloys can be satisfactorily extruded at the relatively slow speeds available when the presses are operated directly from the pumps. On the other hand, many of the newer condenser tube alloys are best made by extrusion, and

further, are best extruded very rapidly, which is possible only by the use of an accumulator.

During the last few years a great deal of attention has been paid to details of mechanical construction, including improved methods of handling so as to speed up the operation and increase production. Such matters are, however, outside the scope of this paper.

Many different copper alloys are extruded commercially and many others are capable of extrusion but have a more limited application. The uses to which these various extruded alloys are ultimately put are very diverse. Certain typical uses are free turning rod, forging rod, valve stem rod, brazing rod, architectural shapes, bolts, rivets, screws, butts, hinges, lock bodies, valve stems, hardware, pinions, gears, welding tips, condenser tubes, distiller tubes, heat exchanger tubes, etc. etc. There are shown in Tables I and II the working and physical characteristics of the alloys most used.

The Muntz metal type alloys containing approximately 60 per cent copper, with or without other elements, were the first to be extruded and still are produced in much greater quantities than all others combined. In particular, free cutting rod having a composition of approximately:— copper 61.5 per cent, zinc 35.0 per cent, lead 3.5 per cent, accounts for a large part of the copper alloy extrusion. This is the standard alloy for high speed screw machine work where the maximum machinability, accompanied by good strength and good corrosion resistance are essential.

Forging brass as the name implies is the alloy most used for the general run of brass forgings. It combines excellent hot working properties with good strength, hardness and good machinability. Various modifications of this alloy are sometimes made but for practically all forging uses such modifications show no improvement over the basic alloy.

Architectural bronze is largely used for door and window cases, store fronts, trim, etc. It is readily hot-worked and can, therefore, be extruded in very complicated sections which would be impossible or impracticable to produce from the other copper-base extrudable alloys. Literally thousands of such shapes are available. Those shown in Fig. 4 will give an approximate idea of the extreme variety regularly produced. On the other hand, due to the relatively low copper content, it is not commercially feasible to cold draw this alloy so that its use is restricted practically to sections in the condition as extruded only.

Table I

Copper-Base Extruded Alloys—Rod

Alloy	Approximate Composition*						Approx. Temp. Range For Extrusion Deg. F.	Approximate△ Mechanical Prop. As Extruded		Most Commonly Used Temper	Approximate△ Mechanical Prop. For Such Temper	
	Cu	Pb	Sn	Ni	Mn	Al	Misc.		Tensile Str. Lbs./Sq.In.	Elong. % In 2"		Tensile Str. Lbs./Sq.In.	Elong. % In 2"
Free Cutting Brass	61.5	3.5						1300-1400	48,000	45	Drawn	55,000	30
Forging Brass	60.	1.75						1250-1350	50,000	45	As extruded	50,000	45
Architectural Bronze	58.	3.						1200-1300	54,000	45	As extruded	54,000	45
Naval Brass	60.		0.75					1200-1350	55,000	45	(Light Drawn)	63,000	35
											(Hard Drawn)	70,000	25
Free Cutting Naval Brass	60.	1.5	0.75					1200-1350	55,000	45	(Light Drawn)	63,000	35
											(Hard Drawn)	70,000	25
Manganese Bronze	57.5		1.		0.05		1 Fe.	1150-1300	65,000	35	Drawn	80,000	20
Muntz Metal	59.							1250-1350	52,000	45	Drawn	62,000	30
Rivet Metal	63.							1300-1400	46,000	50	(Rivet Temper)	60,000	35
											(Hard Drawn)	80,000	10
Free Cutting Commercial Bronze	89.	2.						1400-1500	36,000	50	Drawn	45,000	30
Copper⊡	100.							1500-1650	32,000	40	Drawn	42,000	25
Leaded Copper	99.	1.						1500-1650	32,000	40	Drawn	42,000	25
Silicon Bronze⊙	96.						3 Si.	1450-1550	53,000	70	(Drawn-half hard)	73,000	45
											(Drawn hard)	90,000	35
											(Drawn extra hard)	108,000	25
10% Nickel-Silver	45.	2.		10	0.5			1300-1400	80,000	10	As extruded	80,000	10
13% Nickel-Silver	41.	1.25		13	0.5			1300-1400	90,000	7	As extruded	90,000	7
Nickel-Aluminum Bronze	91.			7.5		1.5		1400-1500	50,000	45	Quenched & aged	100,000	20
Nickel-Aluminum Bronze	85.			12.5		2.5		1600-1700	60,000	35	Quenched & aged	125,000	10
Aluminum Bronze 5%	95.					5.		1500-1600	50,000	60	Drawn	70,000	30
Aluminum Bronze 8%	92.					8.		1350-1450	60,000	60	As extruded	60,000	60

△Properties shown are *average;* not to be used as minimum requirements.
⊙Analysis is for Olympic Bronze—Properties typical of all 3% silicon bronzes.
*Balance in all cases is zinc. ⊡Includes all types of copper.

Table II

Copper-Base Extruded Alloys—Tube

Alloy	Approximate Composition					Approx. Temp. Range For Extrusion Deg. F.	Approximate* Mechanical Properties As Extruded		After Final Condenser Tube Anneal	
	Cu	Ni	Al	Sn	Pb	Zn		Tensile Strength Lbs./Sq.In.	Elong. % In 2"	Tensile Strength Lbs./Sq.In.	Elong. % In 2"
High Brass	66.5				0.5	33	1225-1275	45,000	65	50,000	55
Admiralty Brass	70.			1		29	1200-1250	46,000	65	52,000	60
Red Brass	85.					15	1425-1475	38,000	50	44,000	42
Aluminum Brass	76.		2			22	1350-1400	48,000	65	52,000	60
Ni Silver-18%	65.	18				17	1600-1650	55,000	45	60,000	40
20% Cupro-Nickel	75.	20				5	1800-1850	48,000	40	53,000	35
30% Cupro-Nickel	70.	30					1850-1900	54,000	35	63,000	30
Silicon Bronze	96.		3 Si			1	1350-1400	53,000	65	60,000	60
Nickel-Aluminum Bronze	92.	4	4				1500-1550	50,000	45	60,000	30
Aluminum Bronze	95.		5				1475-1525	50,000	55	58,000	35

*Properties shown are *average*, *not* to be used as minimum requirements.

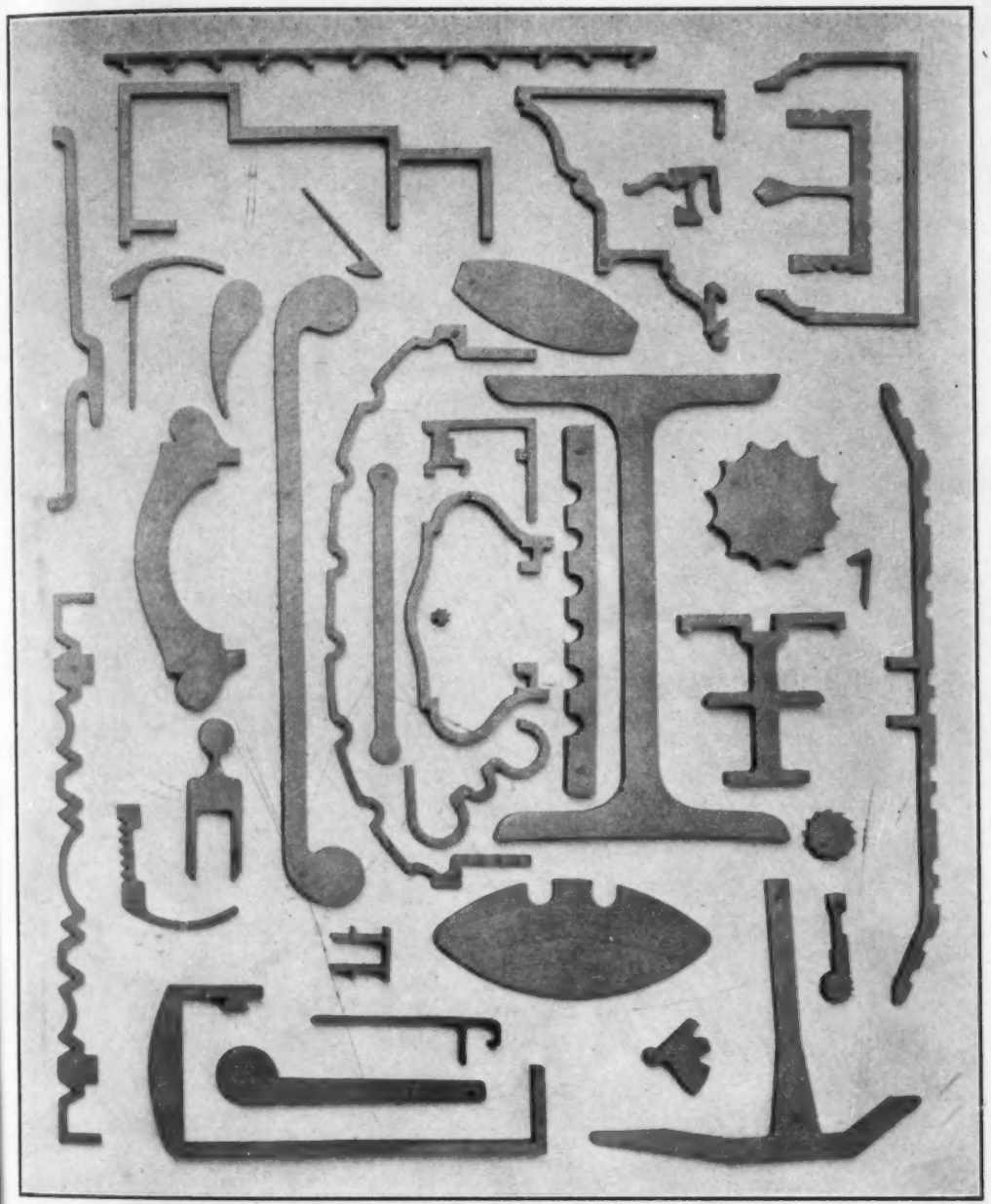

Fig. 4—Photograph Showing Some of the Many Different Sections that Can Be Extruded in Copper Alloys.

Naval brass has long been used for various structural parts on vessels where a strong alloy of good corrosion resistance is necessary. This alloy is readily extruded or forged and in addition can be drawn fairly hard to get high strength and hardness. The free cutting Naval brass is the same alloy but containing lead to obtain improved machinability. This improvement is attained with but very slight

sacrifice of working properties and for many uses it is a more logical alloy than the non-lead bearing variety.

Manganese bronze is quite similar in characteristics to Naval brass but is intrinsically a somewhat harder and stronger alloy and is used in preference to Naval brass where such greater strength and hardness is desired.

Muntz metal is used mostly in the form of brazing rod but is also used to some extent for structural purposes where similar alloys containing lead would be objectionable.

The rivet metal alloy has a sufficiently high copper content to permit of suitable cold working in addition to its capacity for being extruded. It is suitable for general use in rivets, screws, etc.

The free cutting commercial bronze alloy is largely used for hardware and similar uses where a reddish or bronze color, in addition to good strength, workability, machinability and excellent corrosion resistance is desired.

Copper is quite difficult to extrude and there is relatively little production of this metal by extrusion methods. The only justification for extruded copper sections occurs where high electrical conductivity is desirable in addition to other properties obtainable in various extruded shapes. The leaded copper is practically limited to medium size round and hexagon rods from which acetylene welding tips and similar articles are made. The requirements are reasonably good machinability together with the maximum thermal conductivity.

There are many silicon bronzes on the market but the most commonly used ones all contain approximately 3 per cent silicon with or without smaller amounts of any or several of the elements zinc, manganese, aluminum, iron, tin, etc., and the balance copper. Regardless of the exact composition, the working and physical properties are determined almost solely by the silicon content. They have as a group the widest variety of commercial applications of any of the wrought copper alloys. They can be produced in practically all forms and in the form of rod either extruded or hot-rolled and cold drawn are particularly well adapted for high strength corrosion resistant structural parts. A very considerable amount of these alloys is now being used for bolts especially in outdoor structures and in connection with electrical transmission lines.

The extruded nickel-silvers are used mainly for their white color, high strength and fair machinability. While they can be extruded quite readily, the permissible temperature range is narrow compared

to the other extruded alloys and it is relatively difficult to extrude them in small sizes or very complicated sections.

The nickel-aluminum bronzes listed in Table I being precipitation hardening type alloys, are well suited to such uses as line shafting, propeller shafting, high strength forgings, bolts, etc. The particular advantage of these alloys as compared to some others is that they can be worked in the quenched or softened condition and subsequently hardened by aging to give high strength and hardness.

The two aluminum bronzes are fairly readily hot-worked but relatively difficult to cold work. They are used where a high strength, hard, very corrosion resistant material readily adapted to hot working is desired. In particular, these have a very pleasing golden color and are frequently used where this color is the controlling feature.

The tube alloys listed in Table II are largely used either for pipe or condenser tubes. The first three are used in large quantities for pipe, the red brass being the best all around commercial alloy for plumbing use where very corrosive waters are encountered. All of the alloys are used more or less for condenser tubes. The exact conditions under which each is used and their relative merits under various service conditions are outside the scope of this paper. It suffices to say that the commercially more important ones are Admiralty, red brass, aluminum brass and cupro-nickel.

The linear speed of extrusion in rod presses varies enormously with the composition of the alloys, the size and shape of the section, the temperature and the presence or absence of an accumulator. In rod presses, speeds all the way from approximately 20 to well over 500 feet per minute are commercial. In tube extrusion presses using an accumulator, the linear speed is in some instances as high as approximately 1000 feet per minute.

The pressure required to extrude also varies quite widely depending on much the same factors as those which control speed. With the softer alloys, larger sizes, and simpler shapes, pressures as low as approximately 30,000 pounds per square inch might be used. With harder alloys, more complex shapes and relatively smaller sections, pressures up to 120,000 pounds per square inch are found necessary. In general, when the press is operated direct from the pumps, the pressure required to start flow through the die is fairly high. Immediately after starting, the pressure falls to a fairly low figure which holds relatively constant until near the end of the extrusion when the pressure may rise to at least double that required for

starting. This effect, as well as the marked effect of temperature, are both shown in Fig. 5, which shows results obtained in experimental extrusion of free cutting rod through a two-hole die on 0.650-inch diameter.

In ordinary alloys used for structural purposes, most rods are extruded straight, on to either a stationary or traveling cooling bed and are subsequently drawn to finish. Any material which is ultimately to be drawn into rather small diameter rods or wire is normally extruded in sizes from about 3/8 to 9/16 inch, depending on

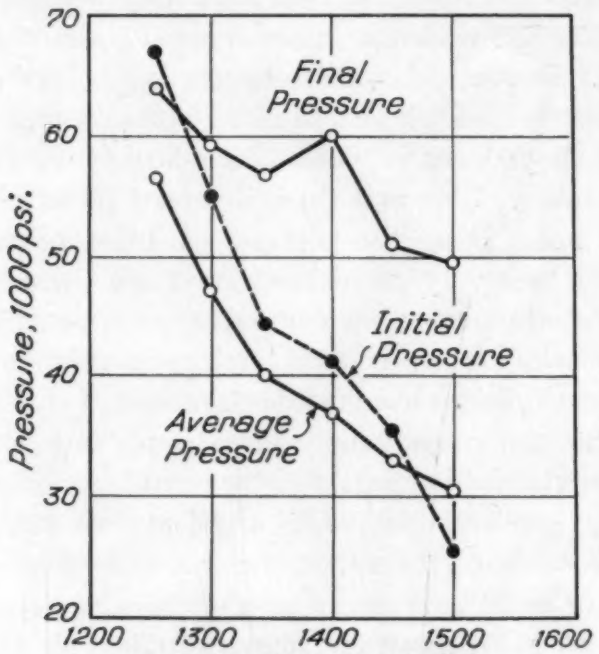

Fig. 5—Effect of Temperature and Pressure in Extruding Free Cutting Copper Alloy Through a Two-Hole Die 0.65 Inch in Diameter.

the alloy, and coiled hot as it comes from the die. In these smaller sizes particularly, several rods may be extruded from a die, sometimes as many as four. Two-hole dies are commonly used up to about 1 inch but larger rods are invariably extruded through a single hole die.

Considerable has been written about the extrusion defect or "pipe" as it is called in the mill. This defect is more or less inherent in the extrusion of brass rods and is caused primarily by the flow of the oxidized surface of the billet to the center during the extrusion process. Many schemes have been proposed and tried and some

are in actual operation which more or less effectively overcome this difficulty. The so-called inverted process was much talked of a few years back and this did accomplish considerable in this direction. In this process a hollow ram is used, and the die fastened to the end of the ram is pushed into the stationary billet so that there is no relative movement between the billet itself and the wall of the container. The objections to this process are that it is very slow and considerable scrap is encountered in the sleeve, which of necessity remains in the container. Much difficulty is found in removing this sleeve, which still further slows production.

In some cases the ordinary direct process has been used, but with a dummy block considerably smaller than the bore of the container so that a sleeve is left and the amount of oxidized material carried into the extruded rod considerably reduced. Which of these methods is the better is perhaps a controversial matter, but the consensus of opinion seems to be that the direct process without the use of a sleeve is so much simpler, quicker and cheaper to operate that it more than offsets the saving due to less scrap from the extrusion defect.

In discussing the extrusion defect, attention is called to the fact that in general the higher copper content alloys having a single alpha solid solution structure behave quite differently. Here little or no central pipe is incurred, but prevailingly the oxidized material comes to the surface of the extruded rod, frequently giving the appearance of a shag bark hickory tree. In such instances the defective surface has to be removed by scalping before the rods can be drawn to final size. Even with this extra operation, however, it is a very satisfactory and commercially feasible procedure.

Hot Extrusion of Aluminum-Base Alloys

The alloys of aluminum are extruded in very large quantity and in a wide variety of forms. The equipment used is quite similar to that employed for the copper alloys. The presses are usually horizontal ones with capacities ranging from about 500 to 4000 tons.

The same general range of size and shape of section are available as with the copper alloys. The shapes shown in Fig. 4, for instance, give a good idea of the variety of shapes regularly produced. In addition, of course, there are the usual rounds and other simple rod sections and a wide range of tubes.

Although practically all aluminum alloys can be extruded if necessary, only a few are ordinarily employed for the production of extruded products. The commercial designations* and nominal composition of these are shown in Table III.

Table III

Designation	Cu	Si	Mn	Mg	Cr
2S	(99	Min. Aluminum)			
3S			1.25		
51S		1.0		0.6	
53S		0.7		1.25	0.25
A 17S	2.5			0.3	
17S	4.0		0.5	0.5	
24S	4.2		0.5	1.5	

The relative ease of extrusion of these alloys is approximately in the order listed, the first being the most readily worked.

Alloys 2S and 3S are subject to strain hardening only and can, therefore, be produced only in the annealed or as extruded tempers. The remaining alloys can be produced in the same two tempers and in addition, heat treated tempers. The properties of the several alloys in their available tempers are shown in Table IV. In all cases the temperature ranging from about 700 to 950 degrees Fahr. is used

Table IV
Typical Mechanical Properties of Extruded Aluminum Alloys (1)

Alloy and Temper	Tensile Strength Lbs./Sq.In.	Yield Strength (2) Lbs./Sq.In.	Elongation (3) % in 2"	Brinell Hardness 500/10	Shear Strength (4) Lbs./Sq.In.	Endurance Limit (5) Lbs./Sq.In.
2S-O	13000	4000	45	23	9500	5000
2S-X (6)	15000	13000	25	28	10000	6000
3S-O	16000	5000	40	28	11000	7000
3S-X (6)	18000	15000	20	35	12000	8000
17S-T	58000	35000	22	100	35000	15000
A17S-T	43000	24000	27	70	25000	13500
24S-T	65000	43000	22	105	40000	14500
51S-W	35000	20000	30	64	24000	10500
51S-T	48000	38000	16	95	30000	10500
53S-O	16000	7000	35	26	11000	7500
53S-W	33000	20000	30	65	22000	10000
53S-T	38000	32000	20	80	26000	11000

(1) Properties will vary with size and shape of extruded shape.
(2) Stress which produces a permanent set of 0.2 per cent of the initial gage length.
(3) Values quoted are for 0.505 inch diameter specimens.
(4) Single shear values from double shear tests.
(5) Based on withstanding 500 million cycles of completely reversed stress, using R. R. Moore type of machine and specimen.
(6) X = as extruded.

*A.C.O.A. designations.

and velocities varying from about 5 to 300 feet per minute, depending on the size, shape and alloy.

Even though the annealed and as-extruded tempers of the heat treatable alloys are commercially available, all of these except 53S are generally used only in the heat treated tempers in order to secure the maximum mechanical properties and maximum corrosion resistance. Alloy 53S has essentially the same corrosion resistance in all tempers.

The aluminum alloys are subject to the same extrusion defect as are the brasses. In the case of these alloys, as well as with the brasses, the indirect methods for extrusion have failed to substantiate the advantages which have sometimes been claimed for this indirect process. Present practice is confined entirely to the direct process.

Hot Extrusion of Magnesium-Base Alloys

The magnesium-base alloys are of relatively recent introduction but for certain applications have already gained fairly wide acceptance. Several of the magnesium-base alloys are commercially extruded, using equipment essentially the same as that for copper-base or aluminum-base alloys.

As a class, magnesium alloys are quite hot short as compared to other alloys already dealt with. Therefore, a narrow range of temperature is permissible and the extrusion rate is low as compared to

Table V
Extruded Magnesium Alloys

Alloy	Aluminum	Manganese	Zinc	Magnesium	Approximate Tensile Strength Pounds Per Sq. In.	Approximate Elongation Per Cent In 2 Inches
A	8.0	0.2	...	Remainder	47000	13
F	4.0	0.3	...	Remainder	40000	16
J	6.5	0.2	0.75	Remainder	43000	18
M	...	1.5	...	Remainder	42000	7

that employed in the brasses. Commercially, speeds of approximately 2 to 12 feet per minute are used.

Normally the temperatures are maintained between 600 and 800 degrees Fahr. for any of the magnesium alloys, the higher temperature being best for producing thin sections and the lower temperature for relatively heavier sections. The physical properties are

somewhat affected by temperature, being decidedly better at the low temperature due to greater grain refinement.

In Table V are given the designations,* analyses and chemical properties of four alloys which can be extruded commercially.

Alloy "J" is at present produced in much greater quantities than the other three alloys listed. In most commercial work the temperature and speed of extrusion is such that the pressure on the metal during the extrusion is between 40,000 and 60,000 pounds per square inch. The starting pressure is naturally somewhat higher than this. The pressure also is greatly dependent on the temperature as shown by the following experimental results made on Alloy "F" on a laboratory scale, using billets 1.5 x 2.25 inches and extruding rod 5/8 inch diameter.

Temperature	Extrusion Pressure
300°C	50,000
350	35,000
400	20,000
450	10,000
500	5,000

The magnesium-base alloys while not available in as wide a variety of irregular shapes as the copper and aluminum alloys are normally extruded in a considerable range of sizes of round and other simple forms. In particular, a considerable number of structural sections such as angles, I-beams, channels and special sections for structural use are produced. These alloys particularly are useful where advantage can be taken of their extreme lightness. For parts that have to be stopped and started very rapidly, the low specific gravity of 1.8 typical of these extruded materials is decidedly advantageous. Another advantage is their excellent machinability. Although actual figures are not available, it is claimed that these alloys can be machined faster than any other metal.

The magnesium-base alloys in general are somewhat less resistant to atmospheric corrosion than the aluminum-base alloys and particularly less so than the copper-base alloys. In most outdoor uses it is recommended that they be protected with suitable paints or other surface coatings, but when so protected they are logical structural alloys.

Hot Extrusion of Zinc-Base Alloys

The hot extrusion of rod and shapes from zinc and zinc alloys has not been commercialized to any large extent because of the ex-

*Designations of Dow Chemical Co.

pense of the process. The expense lies in the greater extrusion pressure necessary than with brass and aluminum and the lower extrusion speed caused by the tendency of zinc and zinc alloys to check and crack as the extrusion speed and consequently the temperature build up.

High grade zinc has been extruded at speeds up to 75 feet per minute. The extrusion pressure approached 100,000 pounds per square inch initially. This material had ultimate tensile strengths of 14,500 to 20,400 pounds per square inch and elongations in 2-inch sections of 4.8-26.5 per cent depending upon impurity content.

The zinc alloys containing copper or copper and magnesium show strengths up to 32,000 pounds and elongations in the order of 3-7 per cent but the extrusion speeds decrease to 12 feet per minute on ½-inch diameter rod and the extrusion pressures run up to 120,000 pounds per square inch. In extruding these alloys, the billets must be preheated to 520-570 degrees Fahr. and the container to 660 degrees Fahr. in order to start the extrusion.

Extruded zinc and zinc alloy rods and shapes having the properties mentioned above are suitable for subsequent drawing with attendant increase in the strength and elongation.

Hot Extrusion of High Melting Point Alloys Nickel Alloys and Stainless Steel

For some years considerable experimental work has been carried out on the high melting point alloys but only within about the last year has any commercial application been made. There is at present one installation where stainless steel tubing is being produced in commercial quantities using a 2000 ton press. There is another installation using a 600-ton vertical press now commercially extruding nickel, Monel metal, Inconel and similar alloys. It is perhaps too early to definitely predict whether these alloys will ultimately be extruded in large quantities or not. The basic difficulty is that several of the outstanding properties of these alloys are precisely those which intensify the difficulties of extrusion. Their great mechanical strength at high temperature and considerable resistance to deformation particularly militate against their production by extrusion methods.

Of course, several copper-base alloys containing considerable amounts of nickel are regularly produced in quantities but these are

not looked upon as nickel alloys but are referred to in the section on copper-base alloys.

In the last analysis, successful extrusion of these high melting point alloys comes down to a question of obtaining suitable steels for the extrusion tools. Much research has been devoted to this problem the last few years and considerable improvements have been made. Even so, before these alloys can be economically extruded on a large scale, presses and tools will have to be made available which will stand pressures at least fifty per cent higher than required for copper alloys, along with temperatures greatly in excess of those used for the copper alloys.

Cold Extrusion—Hooker Process

In the Hooker process relatively heavy walled cup shaped blanks are extruded at high speeds, using ordinary crank presses instead of the usual extrusion press. This operation differs essentially from the normal hot extrusion in two important respects. First, the metal is invariably fed into the press at room temperature and there is little, if any, rise in temperature during the process. At least the temperature of the metal remains well below that of incipient recrystallization so that a hard worked structure and corresponding properties result. Second, the process is carried out at relatively high velocities and on relatively small masses of metal. The Hooker process is commercially applied to two classes of products;—first, small arms cartridge cases, and second, small thin walled seamless tubes.

In Fig. 6 there are shown four steps in the making of a small brass cartridge case. The alloy used is high grade 70/30 or cartridge brass. The first two steps comprise a cut-and-draw and a redraw, each performed in the ordinary manner on a plunger press. The third and fourth operations illustrate the application of the Hooker process. The two sectional illustrations show the position of the tools and the metal just at the start of and near the finish of the application of the Hooker process. It will be seen that with the application of pressure by the shoulder of the punch, the metal is made to flow (i.e. is extruded) through the restricted annular space between the punch and the opening in the bottom of the die.

This same procedure with a slight modification is also used extensively for making small copper tubes particularly for use in auto-

mobile radiator manufacture and for production of cups and tubes of aluminum for various purposes. Where tubes are made the blank is cupped in such a manner that the bottom is so thin it is actually or practically pierced. Air is thus admitted during the extrusion operation preventing the collapse of the thin walled tube. This is not necessary with the relatively heavier wall used for cartridge cases.

This process is well adapted in general to small tubular articles in copper or aluminum where a large number of the same size piece are to be made. For instance, in 70/30 brass cartridge cases, this procedure is justified in competition with the ordinary drawing pro-

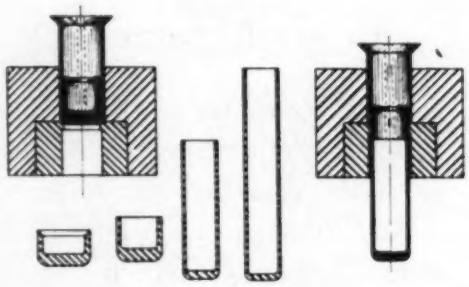

Fig. 6—Diagrammatic Sketch Showing the Steps in Making a Small Brass Cartridge Case by the Hooker Process.

cedure on runs of approximately 10 to 15 million pieces. In the making of these cartridge cases, extremely high tool pressures are encountered, naturally resulting in considerable wear on the dies and punches.

In the manufacture of copper tubes, the sizes which are found economical, run from about 3/16 inch O. D. to approximately ⅜ inch O. D. Some larger diameter tubes have been made experimentally but as yet not in commercial production. The wall thicknesses usually will vary from 0.004 to about 0.010 inch. Anything less than about 0.004 inch is not commercial as it causes excessive tool breakage, and tubes with wall thicknesses over 0.010 inch can be produced at less cost by other methods. The length extruded is usually about 8 to 12 inches and the maximum feasible length about 14 inches.

Although any type of copper probably could be used, oxygen-free copper is the one actually most employed. This, of course, is a homogeneous soft material, well adapted to this operation.

In the sizes ordinarily made the actual velocity of extrusion is much higher than ever attained in the ordinary hot extrusion of either rod or tubes on hydraulic presses. The actual working stroke

consumes but a small fraction of a second and velocities through the die as high as about 2000 feet per minute are attained. This is so rapid that there is little temperature rise in the die and the material itself is heated only slightly. The metal is severely cold-worked and quite high strengths are found in the finished tubes. These may run from about 50,000 to 60,000 pounds per square inch in copper.

The Hooker process also is used in making a considerable range of small seamless aluminum tubes. Like copper, the aluminum is severely cold-worked in this process but the extruded tube may be softened by suitable annealing subsequent to the extrusion.

Impact Extrusion

This process is somewhat similar to the Hooker process but differs from it in some fundamental respects. Whereas in the Hooker process the metal is extruded ahead of the punch through an annular orifice between the die and the punch, the flow of metal in the impact process is in the reverse direction. Here a solid bottom die is used and a flat slug of metal in place of the cup shaped slug of the Hooker process. Here also the punch has no shoulder and when the pressure is applied by the end of the punch to the flat slug, the metal flows back up over the punch. This process is sometimes erroneously referred to as the Hooker process and also sometimes by the misleading term "cold drawing." The term "impact extrusion" seems to be the more logical one to use.

In the making of flat bottom cans or cups of all types, this process frequently competes with similar shells drawn from sheet metal. The relation between the lengths and diameters of the can largely determines which process is the more economical. Relatively shallow articles are cheaper to produce by drawing but with increasing length the advantage swings toward the impact extrusion process.

The impact extrusion process is applied to the making of various aluminum articles including many types of cups or shells, radio tube and transformer shields, grease guns, door check cylinders, flashlight cases, percolator pumps, collapsible tubes and a considerable variety of articles which require some further forming operations after the extrusion.

Shells or cups up to approximately 10 inches in length are found well suited to this process. The diameters may vary ½ inch to about 4 inches, and the wall thicknesses from about 0.003 inch to about

0.045 inch. In general the wall thickness is more or less in proportion to the diameter of the tube or cup.

This process is also applied to making an enormous quantity of collapsible tubes such as used for tooth paste, paint pigments, shaving creams, etc., using pure tin or lead or tin-lead alloys. The pure tin is required for foodstuffs or materials for human consumption, but the lead alloys are found suitable and are naturally less expensive for containers for other types of products.

A great deal of zinc is extruded by this process, using blanks from ¼ to 2-inch diameter and 1/10 to ⅜-inch thick. The diameter of the slug or blank usually equals the outside diameter of the finished cup, and a sufficient thickness is used to provide the desired length at required wall thickness. In the application of the process to zinc, it is desirable to start with metal having a random orientation or at least a structure such that the blank has no directional properties. Otherwise the tops would be uneven, resulting in distortion of the side walls and causing an excessive loss in trimming. It is customary with this metal to heat blanks to a temperature of 390 to 480 degrees Fahr. in order to reduce the pressure required. It has been found that with common zinc a pressure of 87,000 pounds per square inch is required at 300 degrees Fahr. and 40,000 pounds per square inch at 480 degrees Fahr.

Acknowledgment

In the compiling of this paper the author is indebted to many others for information and suggestions concerning the alloys other than those of copper-base. In particular, acknowledgment is hereby made of contributions from the following people:

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N. B. Pilling,
International Nickel Company.
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W. H. Cobb,
United States Rubber Company.
E. H. Dix,
Aluminum Company of America.
W. M. Peirce,
New Jersey Zinc Company.
Roy L. Parsell,
Winchester Repeating Arms Co.

DISCUSSION

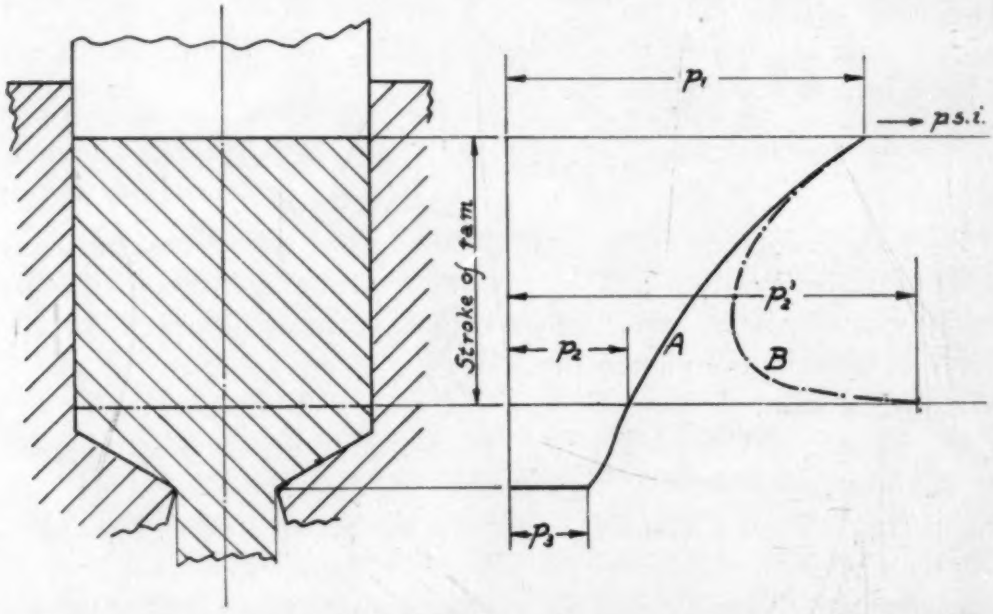
Written Discussion: By C. F. Hammond, metallurgist, Winchester Repeating Arms Co., New Haven, Conn.

In my review of the preprint "Extrusion of Metals" by Mr. Crampton, I find that he has made no mention of lubrication on cold extrusion, by the Hooker process. This is very vital to the success of the operation. For a great many years no other lubricant but lard oil was satisfactory. It is only within recent years that compounded oils have been used. The extrusion presses are piped so that two jets of oil are playing on the extrusion die at all times. It is the proper lubricating and cooling effect of the oils used that keeps the temperature of the extruded article below recrystallization temperature. For later fabrication and final use it is essential in practically all articles we manufacture by this process that the inherent stiffness and hardness from the cold extrusion be retained. The importance of lubrication is therefore apparent.

The entire paper by Dr. Crampton is an excellent contribution to the art of extrusion of metals.

Oral Discussion

E. F. GIBIAN:[1] Those of us who apply the extrusion process to steel, like valve manufacturers, are especially interested in the mechanics of the extrusion process; namely, what happens within the die when the hot metal is extruded. I believe that the copper industry can help us out very much,



because we are using mechanical presses and have no ready means to measure the pressures. In hydraulic presses, the pressure may be readily gaged through the whole cycle of extrusion.

I should like to refer to Mr. Crampton's interesting chart, Fig. 5, where he shows the change from the initial pressure to what he called average

[1]Thompson Products Co., Inc., Cleveland.

pressure and then to the final pressure. Maybe I can best illustrate my point by a sketch. The section to the left represents the general shape of an extrusion die, and the probable distribution of pressure in the metal at the start of the extrusion operation is represented in the diagram to the right by the curve "A". Pressure p_1 occurs under the ram, gradually decreases along the curve "A" and, of course, beyond p_3 the pressure is equal to 0.

It is my belief that the hot metal follows more or less the fundamental laws of hydraulics with certain modifications to allow for the viscosity of hot metal. If we could plot accurately the pressure variation between p_1 and p_3 at the start of the extrusion operation, the resulting curve would have the characteristics of curve "A", i.e. the highest pressure p_1 occurs under the ram, continuously decreases through p_2 to p_3 and then drops to 0. Theoretically the extrusion pressure during the cycle should then follow the pressures of curve "A" from the initial pressure p_1 to the final pressure p_2, but Mr. Crampton's chart, Fig. 5, shows an increase of the final pressure even beyond the initial pressure.

I believe that the high initial pressure is not as much due to the work done on extrusion as to the friction to be overcome on the walls of the die. Then, as the ram advances, the pressure under the ram should change approximately along the curve "A" if the temperature of the hot metal remained constant. But Mr. Crampton pointed out already that the cooling of the material being extruded causes a rise of pressures; towards the end of the cycle the cooling effect is probably considerable due to the small volume of metal and relatively large area of contact with the cooler die walls. Therefore the actual pressure under the ram changes probably in a manner represented by curve "B", which shows only a slight increase of pressures over the theoretical curve "A" at the beginning of the cycle and a very rapid increase at the end of the stroke to the final pressure p_2 in accordance with Mr. Crampton's chart.

I do not know how much research has been done on this problem of ratio of pressures, but it is of vital interest to us who are using extruding dies, because if the pressure required can be reduced we shall increase die life and get better forgings. I should be very much interested to know if Mr. Crampton has any comments.

I. E. WAECHTER:[2] In the extrusion of a 60-40 brass containing in the neighborhood of 3 per cent lead, we are particularly interested in the extruded metal, from the standpoint of the user. In this connection, it was of particular interest to hear Mr. Crampton's remarks that the high temperature extrusions, where it would be easier to extrude, are not visible to us because of the poor structure obtained. I have heard in this day of very high speeds for machining, that it is sometimes necessary to sacrifice quality of the structure itself for the sake of obtaining good machinability at these relatively high speeds. We have found that somewhat larger grain structure does give better machinability. I imagine that higher temperature of extrusion would attain this end. I would like to know what Mr. Crampton's experiences have been.

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[2]Metallurgist, Weatherhead Co., Cleveland.

JOHN A. GANN:[3] Extrusion of magnesium alloys is important from several standpoints. In the first place it is the method now used in commercial production of bars, rods, and shapes of various Dowmetal alloys. Rolling has been tried experimentally on a few simple sections, but at the present time extrusion is the only commercial process. The range of extrusion temperatures is sufficiently broad so that no danger should be encountered due to the temperature becoming too low, and thereby requiring excessive pressure, or due to the temperature becoming too high and, thereby, introducing hot shortness problems. The extrusion process is a practical method of fabricating certain alloys that are too hot short to work by the other methods.

Our second consideration is that extrusion is the first step in the production of forgings and sheet. Cast blanks and billets are rather difficult to break down, but this step is easily accomplished by the extrusion process, and extruded stock forms an excellent starting point for the production of sheet forgings.

The data in Table V could have been enlarged to give some property data corresponding to that in some of the other tables. If that were done, the more important properties that could be included are yield strengths of 27,000 to 34,000 pounds per square inch for the various alloys published, and fatigue endurance limits of 7,000 to 17,000 pounds per square inch. This means that these alloys compare very favorably with the other light alloys.

Lastly, I would like to mention the fact that it is stated on page 74 that data on machinability are not available. Such information is available and has been published in the trade literature, and likewise in one of the earlier volumes of the TRANSACTIONS of the American Society for Metals. In the latter, O. W. Boston showed the superiority of Dowmetal over other metals from the standpoint of machining operations, such as planing, drilling, and milling.

CARL L. JOHNSON:[4] I would like to ask what types of compounding materials are used instead of lard oil, and also are fatty oil emulsions in water used as lubricants in these processes, and if so, how effective are they in comparison with the compound oils. Also are fatty oil and mineral oil emulsions used.

Author's Closure

Mr. Hammond has called attention to the importance of proper and adequate lubrication for the successful operation of the Hooker process. In outlining this paper, all such considerations were intentionally left out in order that the space available be devoted strictly to questions of the general features of the equipment and the characteristics of the alloys used. However, where the lubrication plays so important a part as it does in the Hooker process, reference thereto should most certainly have been included. I appreciate very much that Mr. Hammond has called this to my attention.

Mr. Gibian shows a hypothetical diagram of pressure distribution which is considerably in variance with conditions as indicated in Fig. 5 of the paper.

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[3]Metallurgist, Dow Chemical Co., Midland, Michigan.
[4]Franklin Oil Co., Bedford, Ohio.

It is felt that perhaps there is some confusion as to the interpretation of these data. The pressures shown in Fig. 5 are actual hydrostatic pressures on the metal directly under the ram. In extruding of copper alloys, in ordinary presses it is common experience that a very high pressure is required to start flow. This promptly falls to a relatively low pressure which is maintained throughout most of the stroke and finally climbs to a point even higher than the initial.

At any instant during the extrusion, there might be a slight variation in hydrostatic pressure within the metal from the point in contact with the ram to a high just inside the die, and the shape of this pressure distribution curve might be somewhat as Mr. Gibian has shown. However, in his diagram, pressure p_3 would probably be only very slightly less than pressure p_1 instead of but a small fraction of p_1 as he indicates. Naturally, the pressure does drop to zero as the rod leaves the die, but this drop to zero is very abrupt and the pressure just inside the die not greatly different from that at the face of the ram.

The rise in pressure at the end of the stroke is due largely to the fall in temperature of the then remaining metal and partly to increased internal friction due to the changed configuration of the remaining portion of the billet. It is theoretically possible that an extrusion operation could be carried out so rapidly that the temperature of the billet would not drop and, in fact, might even rise appreciably due to the work done and the lack of time for conduction of heat away from the billet. If this were actually to take place, a drop in pressure at the end of the stroke instead of a rise would be experienced.

Mr. Waechter raises a question as to the correlation between temperature of extrusion, structure and machinability of free-cutting brass rod of about 60 per cent copper and 3 per cent lead. In general it might be stated that in any alloy pressure required for extrusion decreases with increase in temperature. Also, in general, such an increase in temperature results in a coarser crystalline structure which is usually undesirable. In the particular instance in question, an appreciably better and more uniform structure results from low temperature extrusion, which improved structure is reflected in better and more uniform machinability. There is no sacrifice between structure and machinability but the use of a high extrusion temperature would result in a sacrifice of machinability.

Dr. Gann points out the importance of extrusion processes in the commercial fabrication of magnesium base alloys. He gives certain data supplementing those in the paper with regard to yield strength and fatigue endurance limits. The addition of these data to those given in the paper is greatly appreciated.

Mr. Johnson asks what type of lubricants are used in extrusion processes and inquires particularly about fatty oil emulsions in water. In general, in the extrusion of copper alloys, lard oil bases are used, often with the inclusion of fine graphite in suspension. So far as I know, fatty oil emulsions in water are not used. The lubricants favored in the extrusion of alloys other than copper base are not familiar to me personally.

COLD HEADING—BOLTS, RIVETS, NAILS

By Roy H. Smith

Abstract

This paper describes modern cold-heading practice. Single and multi-blow heading methods and their limitations are discussed. Recent developments in multi-die, single stroke machines for heading bolts and nuts are described. Illustrations show many odd and difficult types of cold heading.

COLD HEADING has been defined as the process of forming, by pressure, the heads of bolts, rivets, nails, etc., on the ends of suitable lengths of wire at or near room temperature. Today this definition covers only a small portion of the cold heading field, as there are many articles being produced which bear little or no resemblance to bolts or rivets, as frequently the upset or forming is not done on the end of the wire. Illustrations of such products are shown and discussed later in this paper.

The development of the cold heading industry is due to a number of factors. Formerly, all upsetting was done hot and principally by hand. The author has in his possession some hand-forged nails made many years ago, which are typical examples of this art. Today, practically all hot heading is done in open dies. Comparatively little hot heading is done in the United States in solid dies, although a considerable amount of solid die hot heading is done in Europe. The increasing demand for smoother shanks on bolts, more accurately maintained and free from scale, brought into existence the cold heading industry. The fact that many of the commercial metals and alloys, as now made, are malleable at or near room temperature and are capable of being plastically deformed, has brought about the development of the industry to its present state. While there has been very little information published on the theory of cold heading as such, considerable data are available on the cold working and plastic

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A paper presented as part of the Symposium on the Plastic Working of Metals, Eighteenth Annual Convention of the Society held in Cleveland, October 19 to 23, 1936. The author, Roy H. Smith, is vice president, Lamson and Sessions Co., Kent, Ohio. Manuscript received June 23, 1936.

forming of metals. Some of this information is applicable to the cold heading process.

One of the most complete discussions on the subject is "The Plastic Forming of Metals" by Dr. Siebel as reprinted by *Steel*. Part I of this translation covers "The Laws and Fundamentals of Plastic Forming" from which the following is quoted:

"The capacity for deformation (by compression) becomes greater with increasing average pressure and very great deformations can accordingly be produced under such stress relations.

"As the deformation resistance increases inordinately with increasing deformation during cold working and the capacity of the material for deformation is continuously diminished; it is usual in large cold deformations—to subdivide the process of deformation and to restore the original condition of the material by means of intermediate annealing operations.

"The capacity for deformation of metals is much greater the greater fineness of grain the structure possesses.

"The effect of the velocity of deformation upon the shape of the flow curve in cold working is comparatively insignificant."

Another theoretical presentation is "Plasticity" by Dr. Nadai,[3] which is described as "A Mechanics of the Plastic State of Matter."

Members interested in any field involving the cold- or hot-working of metals might well consult these two authorities.

Other less technical but practical discussions are "Cold Heading" by Lucas and Duston[1] and a later discussion of their paper by the writer.[2] In this discussion it is pointed out that the shape of the cut-off and the fit of the blank in the heading die largely determine the amount of stock which may be upset.

In some phases of the industry the limit of malleability of the raw material has been reached and any future development will depend upon our ability to produce materials possessing greater malleability. As our knowledge of the effects of impurities on the physical properties of metals is increased and methods of controlling these impurities develop, such raw materials may reasonably be expected.

Along with the development of purer raw materials it will be necessary to improve the physical condition of the raw materials with

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[1]"Cold-Heading." By C. L. Lucas and E. W. Duston—*Machinery*, May, 1913—July, 1913—Feb., 1914.

[2]"Cold-Heading of Wire." By R. H. Smith—*Machinery*, October, 1917.

[3]"Plasticity." By Dr. A. Nadai—McGraw-Hill—1931.

respect to surface conditions such as seams, laps, etc. Since the exterior portion of any cold upset is in lateral tension, such surface defects are considerably magnified in the upsetting operation. These defects rarely affect the physical strength of the article, but are objectionable from the standpoint of appearance. See Fig. 1.

In other phases of the cold heading industry, equipment and tools have been the limiting factor. The requirements of mass pro-

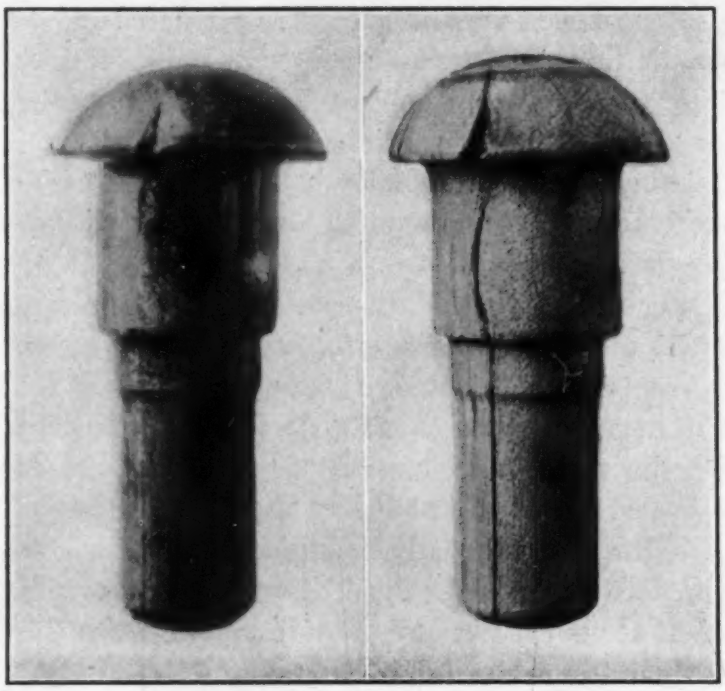

Fig. 1—Showing Surface Defects on Cold Headed Rivets.

duction industries for higher strength material, smaller tolerances and finer finishes have necessitated the development of larger and more rigid heading machines. This in itself has involved no great difficulty other than large capital investment. However, designers have been limited by the ability of the tools to produce a satisfactory product at a satisfactory price. These difficulties have been overcome to a considerable extent by the combined efforts of tool steel manufacturers and die designers, so that today the limits of the cold heading process are more economic than mechanical.

The solution of the heading tool problem has involved the substitution of high alloy tool steels (high speed or high carbon-high chromium) for the 1 per cent carbon tool steel used for so many years. This substitution in turn has made necessary many changes

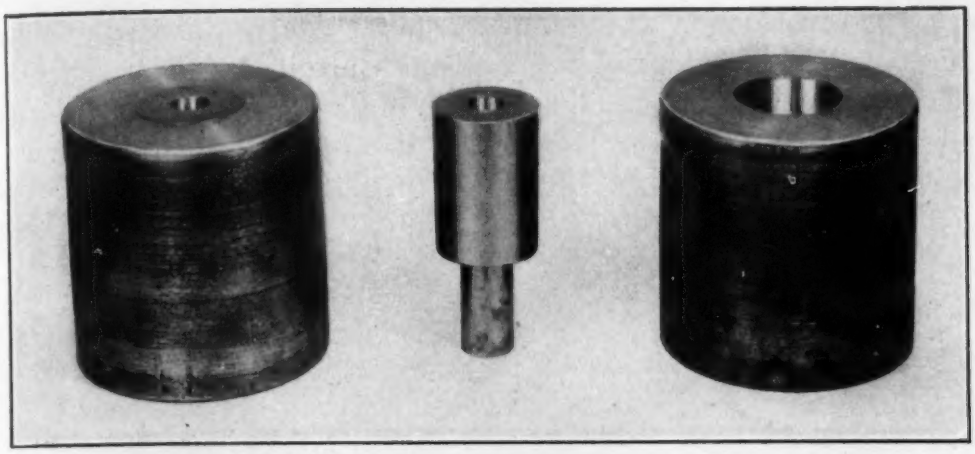

Fig. 2—Photograph Showing Composite or So-Called "Insert" Dies.

in die design. The present trend is toward composite or so-called "Insert" dies, as shown in Fig. 2.

This type die consists of a shell or holder of heat treated alloy machine steel (S.A.E. 3250 or similar material) into which is pressed or shrunk an insert of high alloy steel containing the die impression. By maintaining certain ratios between D_s and D_I (the diameter of

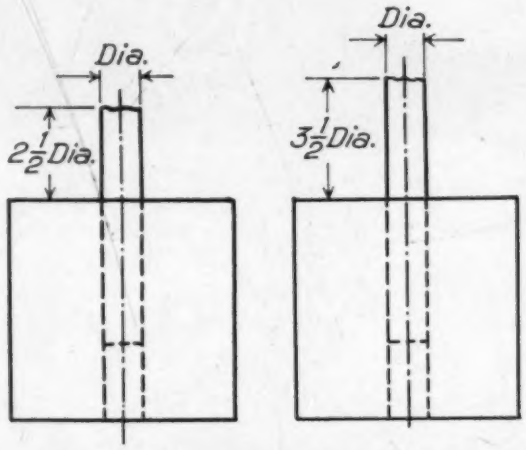

Fig. 3—Diagrammatic Sketch Showing the Amount of Stock Which Can Be Upset in Single Blow Work.

Fig. 4—Diagrammatic Sketch Showing the Amount of Stock Which Can Be Upset in Double Blow Work.

shell and insert respectively) and a definite press fit between the two, it is possible to pre-load the insert to such an extent that it is always in compression. By this method, the abrasion and wear resisting qualities of the insert are used to the fullest extent, together with the strength and toughness of the alloy machine steel.

The construction of these dies requires fine workmanship and a thorough understanding of the mechanics involved. However, they make possible the cold heading of very difficult types of work. By their use it is commercially practical to make bolts, etc., with exceedingly close tolerances.

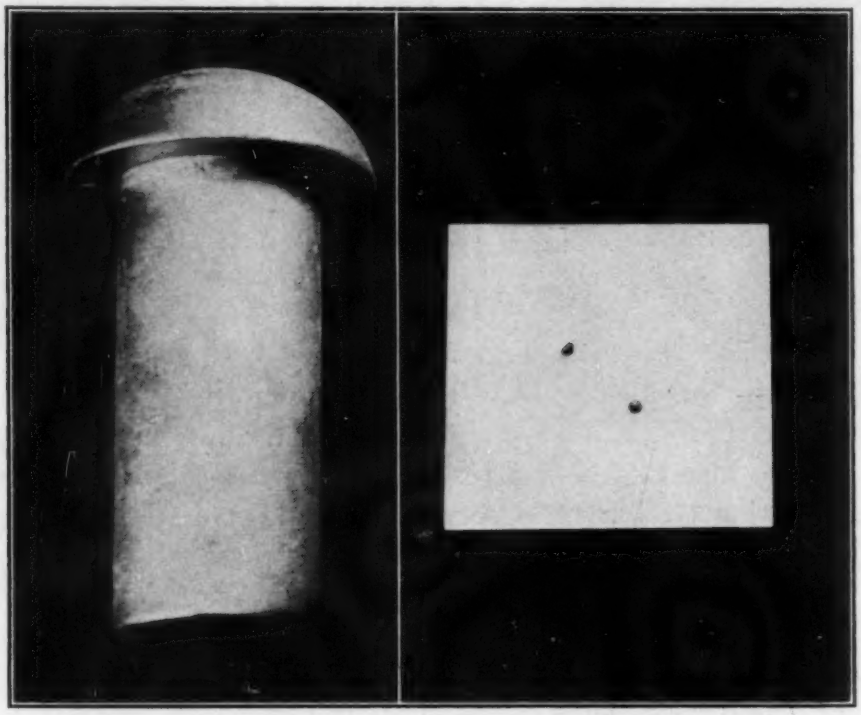

Fig. 5—Photograph of Two Rivets, One Near the Lower Range in Size and the Other Near the Upper Range. Note That it Was Necessary to Place the Smallest Rivets on a White Background in Order to Show Their Size.

For many years the design of cold-headers has followed certain conventional lines. In these machines, the amount of stock which can be upset is much less than the theoretical limit, due to the cut-off, curvature of blank, eccentricity of loading, etc. It has been found by experience, that the unsupported end to be upset should not exceed 2½ wire diameters for single blow work, nor 3½ wire diameters for double blow work. With the best of practice, these limits can and are being exceeded, but in a general sense they hold true. See Figs. 3 and 4. Fig. 5 illustrates two rivets, one near the lower present range in size and the other near the upper. It was necessary to use a white paper background to show the smaller rivets and even then they look like specks of dust.

In recent years several processes and machines have been devel-

oped which make possible the collecting of greater amounts of stock in the head section. In the Kaufman single extrusion process,[4] wire of nominal diameter (instead of pitch diameter) is pushed into the solid heading die and reduced to pitch diameter for the thread section. The

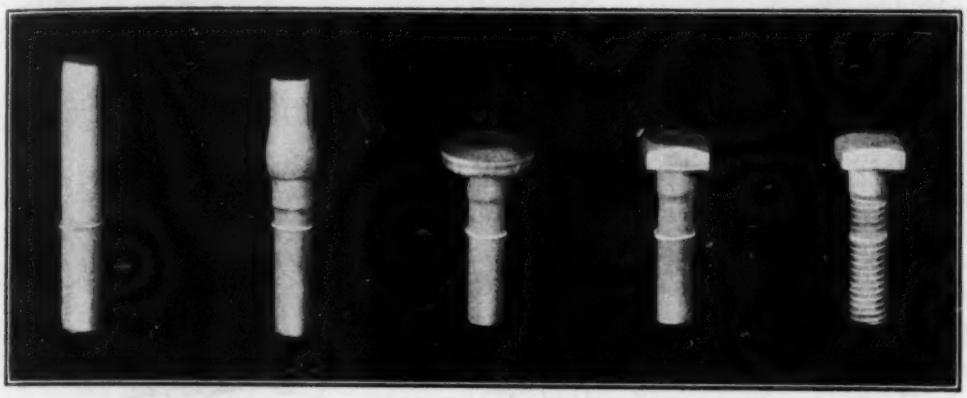

Fig. 6—Illustrates the Steps in the Manufacture of a Square Head Machine Bolt Made by the Kaufman Single Extrusion Process.

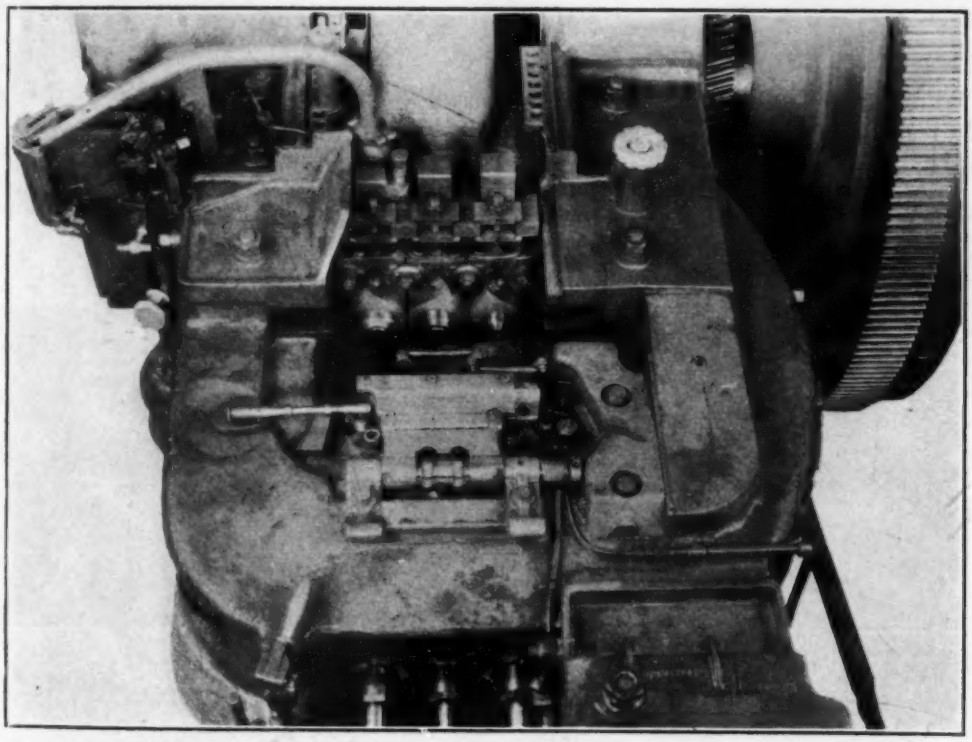

Fig. 7—Photograph Showing a New Type Header Which is Capable of Extending the Range of Cold Heading and is of the Multi-Die Single Stroke Die.

head is then upset in the usual manner. By using wire of nominal size instead of pitch size, a considerable larger volume of stock may

[4]Covered by Kaufman patents.

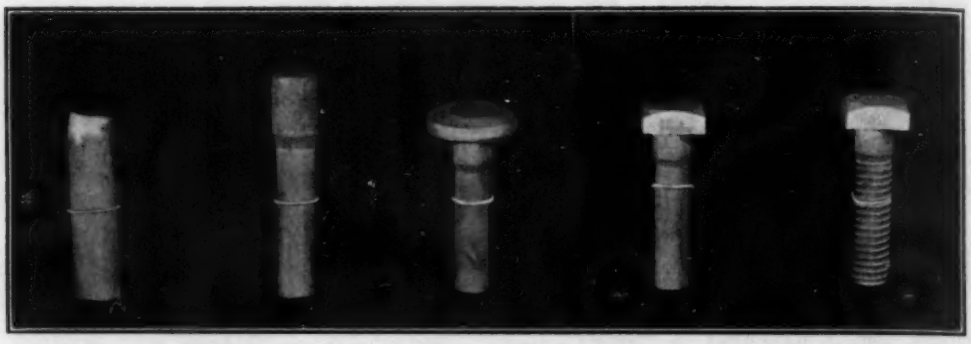

Fig. 8—Photograph Showing the Steps in the Manufacture of a Square Head Machine Bolt by the "Double Extrusion" Process.

be upset, or for a given head size, less plastic flow of the metal is produced. Fig. 6 illustrates the steps in the manufacture of a square head machine bolt by this process.

With the development of the single extrusion process, the single die, double stroke header seems to have reached its limit as regards amount of stock upset. Improvements in squareness and smoothness of cut-off and maintenance of alignment of the heading tools may extend the range slightly, but the percentage increase cannot be large. However, a new type header is available which does extend the range of cold heading. This type, known as the Boltmaker, is of the multi-die, single stroke type. Fig. 7 shows this machine in some detail.

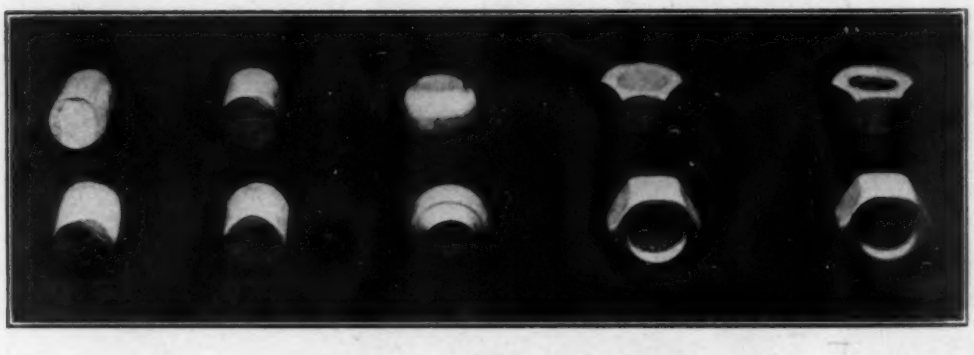

Fig. 9—Photograph Showing the Sequence of Heading Operations in Making a Nut in a "Scrapless" Nut Machine.

It has a wider die space than the ordinary header and a series of dies arranged in a horizontal plane in the bed of the machine, each die being opposed by a heading tool in the slide. Each die is provided with a suitable kick-out pin. The machine is equipped with a transfer mechanism which grips each blank as it is ejected from the die and

shifts it to the next die. In this manner a blank is started, a blank is finished, and intermediate blanks are formed in a new die-hammer combination at each stroke. This type of machine opens up many new fields. One of the first to be developed is the process known as "Double Extrusion",[4] Fig. 8 illustrates the steps in the manufacture of a square head machine bolt by this process.

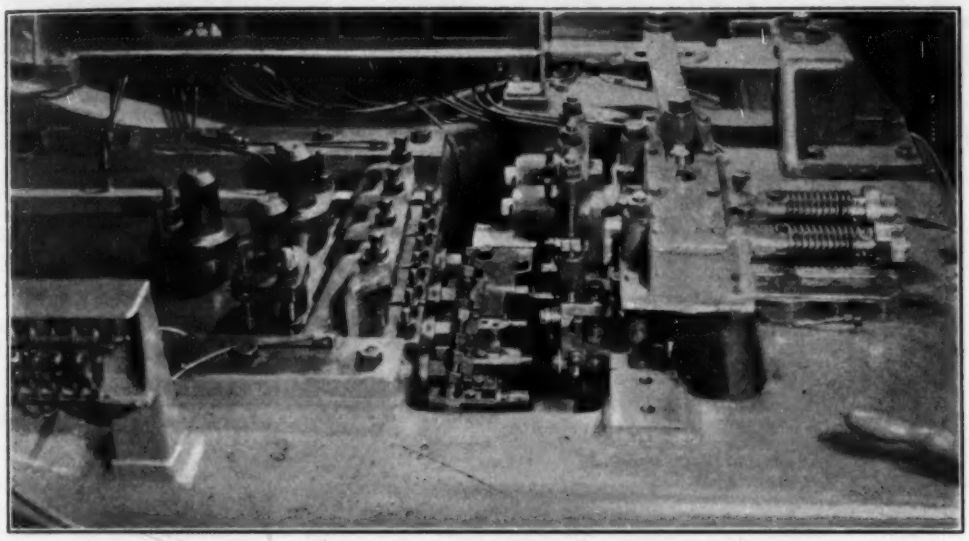

Fig. 10—Photograph of "Scrapless" Nut Machine.

In the double extrusion process the diameter of the wire used is usually $\frac{1}{16}$ inch larger than the unthreaded portion of the bolt, i.e., D + $\frac{1}{16}$ inch. The shank portion is first pushed into an extrusion die and reduced to the nominal diameter D. In the next heading station this shank is again extruded in part—this time to the pitch diameter. The head is upset from the larger material (D + $\frac{1}{16}$ inch) and consequently experiences a minimum of cold-working.

It is not always necessary to heat treat a bolt so made. The heads will seldom give trouble from excessive cold-working strains and the additional cold working of the threaded shank portion gives a higher tensile strength. Where high resistance to fatigue or shock are desirable, suitable heat treating must be done.

One of the most recent developments in the art of cold heading is the so-called "scrapless" nut process, which involves the cold-upsetting of a nut blank from round wire or rod. A blank is sheared from a coil of wire and, in successive stations, formed into a hex-

[4]Covered by Kaufman and other patents.

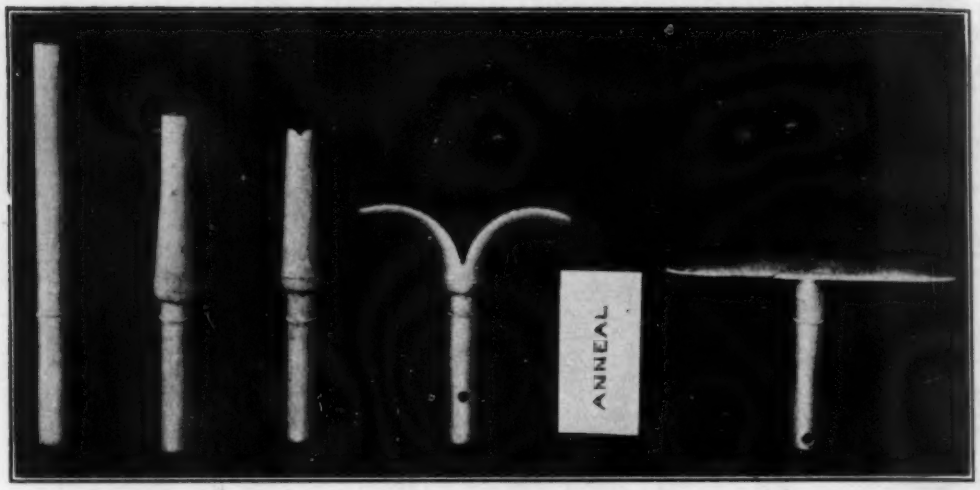

Fig. 11—Photograph Showing the Steps in a Typical Three Blow Operation Job.

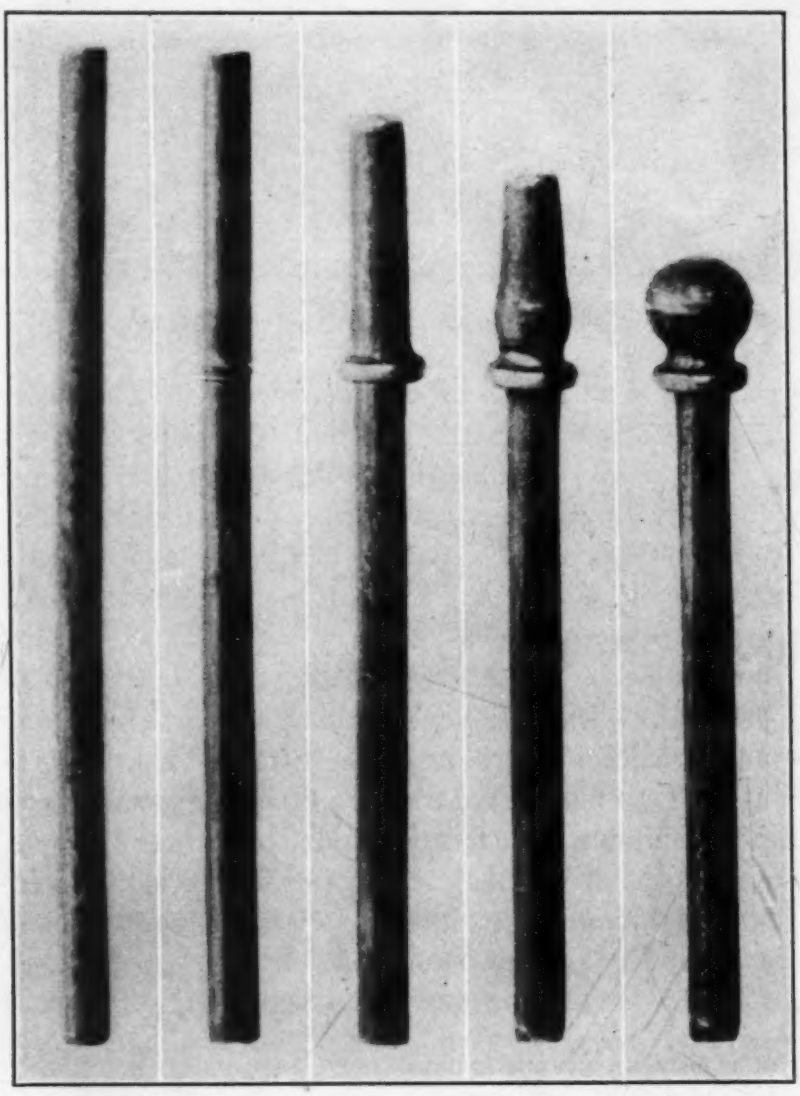

Fig. 12—Photograph Showing the Steps in a Typical Four Blow Operation Job.

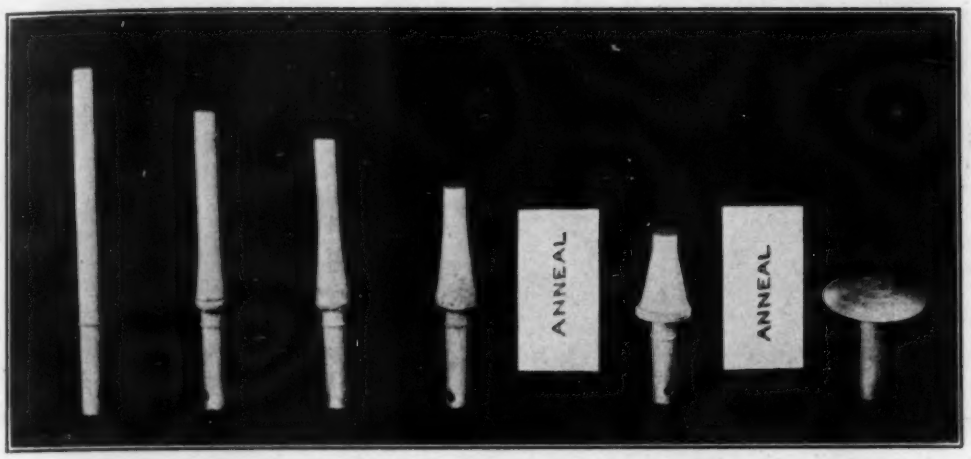

Fig. 13—Photograph Illustrating the Steps in Making a Bow Socket Rivet.

agonal or square nut blank. In the final station, the blank is pierced so that it falls from the machine ready for tapping. The only material lost in this process of cold forming is a slug having a diameter equal to the hole diameter and a thickness approximately one-half that of the finished nut. The sequence of heading operations is shown

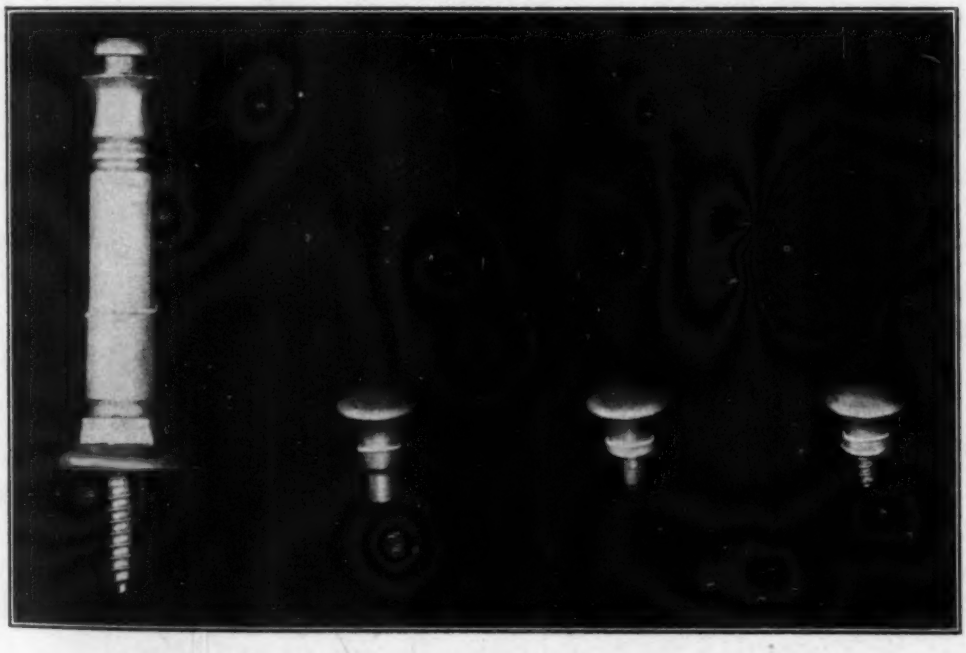

Fig. 14—Photograph Illustrating Pieces Made by a Rough Upset with Subsequent Cutting, Trimming and Pressing to Shape.

in Fig. 9. This "scrapless" nut machine is in reality a multi-die, single stroke header of the solid die type. The machine is illustrated

in Fig. 10. Its success is due to the fact that tool steels are now being produced which will withstand the tremendous pressure necessary for these forming operations.

No discussion of the mechanical processes involved in cold heading would be complete without a reference to products which

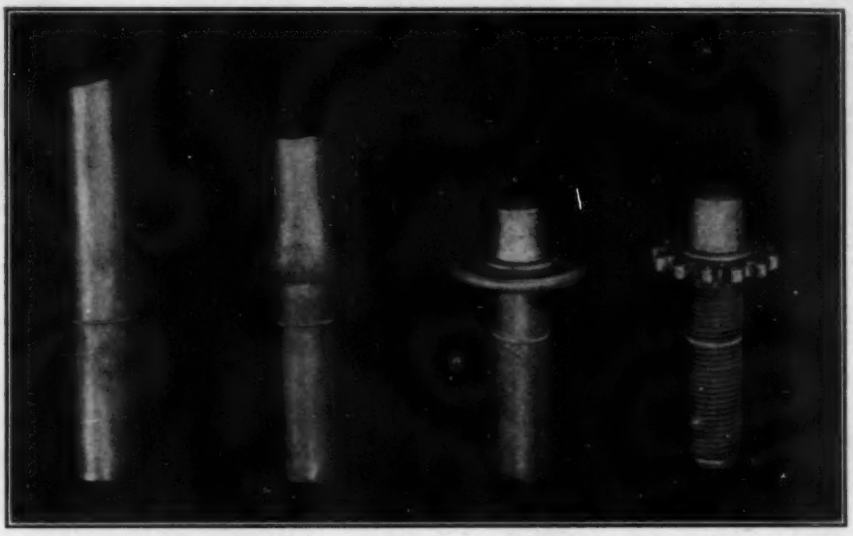

Fig. 15—Photograph Showing an Example of Heading Combined With Later Trimming and Cutting Operations.

bear little resemblance to common bolts, rivets or nails, but which are successfully cold-headed in large volume.

Multiple blow headers have no theoretical limit in length of upset, but practical difficulties place a limit somewhere near to eight or ten diameters, before annealing for reheading or other methods must be used. Fig. 11 shows a typical three-blow job and Fig. 12 a typical four-blow one.

Practical conditions frequently lead to the use of reheading with intermediate annealing in preference to multi-blow headers. Fig. 13 illustrates such practice in a bow socket rivet so commonly used before the days of the closed car body.

Bow socket rivets, California tire rivets and other large heads were also made by swaging larger wire to the necessary shank diameter, then upsetting the larger section in one or two blows.

It should not be overlooked that some jobs can most economically be done by comparatively rough upset and subsequent cutting, trimming or pressing to shape. Fig. 14 illustrates two such items manufactured by Rockford Screw Products Company. One is a finished

Fig. 16—Photograph Showing the Steps in Making Copper Commutator Segments.

door stop, and the other an upset and finished drawer pull. Note the upset at each end, although the holding wire somewhat overshadows this on the drawer pull. Fig. 15 illustrates another example of heading combined with later trimming and cutting operations.

Fig. 16 illustrates an interesting and unusual cold heading job with subsequent trimming, made by the Ford Motor Company. The formed section made of copper is cut off as illustrated—headed two

Fig. 17—Photograph Showing the Steps in Making a Brake Shaft Housing Wherein the Piece is Upset in More Than One Place.

blows and then trimmed to form the commutator segments at the right.

Theoretically, and, to a lesser extent, practically, there is no reason why this amount of material represented by the 2½-diameter

rough rule, should not be upset in more than one location on a blank simultaneously—in fact, in many places.

Fig. 17 is a good illustration of an upset in more than one place. This is the Ford Motor Company's brake shaft housing. After the blank is ejected from the double blow header, it is fed into another header where the two upsets are made simultaneously.

When a very large head is necessary and strength is not the controlling factor, a combination of cold heading and welding is sometimes resorted to.

The foregoing discussion applies equally well to either ferrous or nonferrous materials, but many interesting things are done with nonferrous materials which are not yet done with ferrous.

It is plain that when an upset as illustrated in Fig. 13 is completed, the shank of the material is just about as dead soft as it can be made. Consequently, for work where stiffness or strength of the shank is useful, it is clearly an advantage to change to the method shown in Fig. 6, where, within the limits of the material itself, any degree of stiffness, strength and accuracy can be secured.

Modern cold heading is really a steady march in step with the improved manufacture of cold heading steels and die steels. Many things that are simple today would have been impossible a few years ago and are still impossible where our excellent heading steels and die steels are unavailable.

The present trend in cold heading of bolts, rivets, etc., has been evident for some time. It is toward more and more difficult and unusual types of product. Those products requiring higher and higher unit pressures in the die will probably furnish the most interesting developments in the immediate future. Extrusion in combination with upsetting and pressing, and possibly with some slight elevation in temperature, will play a major part.

Acknowledgments

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Oral Discussion

I. E. WAECHTER:[1] How does the finish of the material effect the life of the dies, and what hardness of the die insert is permissible to give a good die life, and what hardness of the trimming die is permissible to obtain good die life? How much finish must be allowed to be removed after hardening, in order not to try to work the trimming die with a soft skin if we do not remove adequate amount of finish?

R. J. HAIGIS:[2] I would like to get information relative to the type of steel used in the tools which cut off the slug that is carried over to the dies for heading. We have found that instead of cutting off clean, low carbon wire has a tendency to mush rather than cut off, and the squareness of the slug as it is carried over to the dies makes a lot of difference. We have had some difficulty in getting the proper type of steel to stand up for this cut-off.

Author's Reply (by C. L. Harvey)

With reference to the first question asked as to the finish of raw materials for bolts and nuts, for the ordinary extruded bolt by the single extrusion process, we used a cold drawn wire which has been given a moderate coating of lime after the pickling operation, and has then been drawn through a lime-aluminum stearate mixture. The success of the wire in the single extrusion process seems to be the amount of lime used on the raw material.

As far as the stearate lime mixture in the wire drawing box is concerned, there are other compounds on the market that will give satisfactory results.

In the double extrusion processes, we use a much heavier coating of lime. As a general rule, our raw material is sub-coated after it is pickled, and then heavily limed and finally drawn through lime-aluminum stearate in the block.

The finish on this raw material does play an important part in the die life. If the coatings are too light, it is impossible to get a satisfactory die life. The die will cut out after a short run. As to the hardness of the insert dies, our practice is to use a standard 18-4-1 high speed steel of approximately 0.70 per cent carbon for the insert, heat to 2280 degrees Fahr., oil quench, and then draw back to a Rockwell hardness of C-60. There are certain instances where the die can be as hard as Rockwell C-62, and there are other types of work where it is drawn to Rockwell C-58, but for general types of production a Rockwell C-60 seems to be about right.

In regard to trimming dies, our practice is to use a standard 18-4-1 steel of somewhat lower carbon content (approximately 0.55 to 0.60 per cent) and heat treat to a Rockwell C hardness of 57 or 58.

As far as grinding allowance for trimmer dies is concerned, we do not have any. I believe that the proper thing there is to use a high speed furnace of such construction that a soft skin is not obtained on the die.

With reference to cut-off tools for cold headers, we use several types of steel. One of them is a straight carbon steel of approximately 1 per cent carbon, water hardened and tempered to Rockwell C-58. There are certain instances where we are able to use high speed steel, but that is the exception rather than the rule.

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[1]Metallurgist, Weatherhead Co., Cleveland.

[2]Metallurgist, Stanley Works, New Britain, Conn.

COLD FORMING PROCESSES—DRAWING RODS AND BARS

By J. E. Beck

Abstract

This paper deals with the general aspects of cold-drawing rods and bars in the so-called plain carbon steel grades. Considerable space is devoted to a description of the cold drawing process itself. While processing practices differ to some extent, such differences usually exist in the details of handling which, while apparently unimportant to the casual observer, nevertheless assume great importance in the deliberate production of specific results. This is particularly true with regard to the control of physical characteristics, machinability, finish, and in controlling size accuracy, uniformity of section and a high degree of straightness.

IT would seem to be rather common practice among steel users to describe any type of cold-finished bar product merely as cold-rolled or cold-finished bars. Both terms are broad and indefinite and do not designate the specific method of cold processing employed, of which there are several. The more common methods, descriptively named are cold-rolled, cold-drawn, cold-drawn and ground, turned and polished, and turned and ground. In Table I are shown the several types of cold-finished materials mentioned with a brief description of their respective physical characteristics, types of finish and degrees of size accuracy.

All cold-finished material, regardless of the method of processing, is preferred over hot-rolled for one or more of the following reasons: (a) Special size accuracy; (b) smooth or bright finish; (c) characteristic physical properties; (d) better machinability. Of the five production methods mentioned cold drawing and turning and polishing are most popular, with the former occupying the foremost position. This is true because cold drawing is applicable to a wide range of shapes including rounds, hexagons, squares, flats and special

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Table I
Types of Cold-Finished Steel

Processing	Physical Characteristics	Description
Cold-Rolled	Cold-Drawn Properties	Commercial cold-drawn finish and size accuracy
Cold-Drawn	Cold-Drawn Properties	Commercial cold-drawn finish and size accuracy
Cold-Drawn and Ground	Cold-Drawn Properties	Ground finish with very close size accuracy
Turned and Polished	Hot-Rolled Properties	Typical turned and polished finish, but same size variation as cold-drawn
Turned and Ground	Hot-Rolled Properties	Ground finish with very close size accuracy

sections not to mention the characteristic physical properties and improved machining quality resulting from this practice.

It is not the purpose at this time to discuss any of these methods other than cold drawing although in view of the five practices employed commercially in the production of cold-finished material, failure to make some reference to these methods and draw special attention to the fact that this paper will deal with cold drawing only, might prompt some confusion.

Historical

The practice of cold working metal to produce wire is an ancient one. The Bible records one incident of wire production in the third verse of the thirty-ninth chapter of Exodus which states, "and they did beat the gold into thin plates and cut it into wires—." Obviously, hammered wire so produced would be flat. While there is little definite information regarding wire production as far back as this, it is not likely that cold drawing through a plate or die block had been developed. However, it is a matter of record that the die block method was in operation as early as the Eighth Century, and it is well known that wire drawing on a commercial scale was being practiced in Europe as early as the Thirteenth Century.

The first wire drawn in this country was done in New England in 1775. Since the open-hearth and Bessemer steel making processes did not make their appearance before 1858, the materials used for wire in this country between 1775 and about 1860 were wrought iron, copper, brass, bronze, etc. However, with the rapid adoption of steel, wire drawing kept apace of steel production developments after 1860.

It was at about this time that Bernard Lauth patented a cold rolling process for wrought iron. By this method, rounds larger than wire sizes were cold-finished by passing a hot-rolled bar, cleaned of

scale, repeatedly through rolls, effecting a light reduction in each pass, until the desired size was obtained. The resultant product, generally speaking, was quite similar to the present cold-drawn bar product in appearance, finish, size accuracy and physical characteristics. With the introduction of steel, cold rolling of shafting for power transmission purposes, machinery and agricultural equipment became quite popular.

As a commercial production method, cold rolling is a comparatively slow and cumbersome process. In 1900 or about then, after the similarity in the general characteristics of cold-rolled shafting and cold-drawn wire products became fully recognized, it followed that producers of cold-rolled shafting developed cold drawing methods, patterned after wire-drawing practices, to accommodate bar sizes. Today there is very little cold-finished material actually cold-rolled. Such cold-rolling as is done is confined almost exclusively to specialties which do not lend themselves easily or economically to cold drawing such as wide thin flats, special sections, small nut flats, etc.

Cold Drawing Procedure

Cold-drawn material is produced or drawn from hot-rolled bars or coils prepared especially for the purpose. As the name implies, cold drawing involves primarily a stretching or elongating action which accentuates hot-rolled surface irregularities such as scabs, slivers, seams, pits, and scale pitting. Further, the internal character of the hot-rolled bar must be of a high order because pronounced chemical segregation, porosity and pipe might cause objectionable difficulties such as size and section irregularities or breakage in drawing. Such quality irregularities, particularly the surface defects, might also damage the drawing dies themselves. Apart from production troubles apt to be encountered as a result of quality irregularity, the applications of cold-drawn steels are quite severe and a high standard of quality must be delivered if complaint is to be avoided. Consequently, since it is the keystone of cold-drawn quality, so to speak, the hot-rolled bar must be of very high caliber generally if an acceptable cold-drawn product is expected.

Regarding the cold drawing practice itself, the order of operations follows: (a) pickling; (b) washing with water and immersing in lime solution; (c) pointing; (d) cold drawing; (e) straightening; (f) cutting to length; (g) inspecting and shipping. Each of these steps will be discussed in some detail.

Pickling—The purpose of pickling is to remove the hot rolling scale which if allowed to remain on the bar would mar the cold-drawn finish, either because of its unsightly presence on the bar after drawing, or because particles would clog the die and scratch the bar during the drawing operation. Further, scale might also damage the expensive dies themselves thereby reducing die life.

Pickling consists of immersing the hot-rolled product to be cold-drawn in a dilute solution of sulphuric acid, varying from 3 to 8 per cent, to which an inhibitor has been added. An important factor in pickling is the temperature of the bath. At atmospheric temperatures, the action between acid and metal would be comparatively sluggish. On the other hand, at 190 or 200 degrees Fahr., the acceleration resulting from the higher temperature would be accompanied by heavy losses of acid due to evaporation. Consequently, it is more or less common practice to employ a temperature in the neighborhood of 150 degrees Fahr.

The operating action of the acid in removing scale is largely mechanical. The chief constituent of the scale, ferroso-ferric oxide, Fe_3O_4, is only slightly soluble in dilute sulphuric acid under pickling conditions. The pickling solution, however, does penetrate the coating of scale and attacks the steel itself underneath forming iron sulphate with a vigorous evolution of hydrogen, principally. The hydrogen thus liberated literally blasts or blows the scale from the bar after which the loose particles settle in the bottom of the pickling vat or tank necessitating removal periodically.

The length of time required to pickle satisfactorily depends upon many variables such as size of the bar, chemical analysis of the steel, acidity and condition of the pickling bath, character of the scale, temperature and possibly other factors. While those skilled in the practice of pickling could predict with a fair degree of accuracy the pickling time required under known conditions, in the final analysis, the job rests principally upon the judgment of those handling this work. Obviously, a poorly handled pickling operation might ruin the stock for subsequent operations. If not pickled sufficiently, some scale might remain which would be objectionable. On the other hand, "over-pickling" must be avoided so as to minimize "acid pitting" and mottled areas which would carry through to the finished product. Acid brittleness, an ailment which prompts some concern in high carbon wire, causes little or no serious trouble in the bar sizes, which are appreciably larger than wire in cross section area.

Washing and Liming—After proper pickling, the material is removed from the pickling vats and is washed as thoroughly as possible with a water stream. The water pressure washes off particles of loose scale and also minimizes and dilutes further any free acid present. It would be expecting too much to expect to free the material completely of acid by such washing alone and obviously any free acid remaining would cause subsequent pitting and corrosion, both of which would be objectionable in a cold-drawn product. Consequently the material, after washing, is dipped into a slaked lime emulsion which serves a two-fold purpose. First, any free acid present is neutralized; and second, the lime coating which dries on the bars acts as a lubricant thereby preventing die scratching in the subsequent cold drawing operation. The lime bath, whose consistency is regulated in keeping with the thickness of coating desired, is usually steam heated to a temperature of approximately 175 degrees Fahr.

Pointing and Cold Drawing—The actual drawing operation consists of pulling or drawing the pickled and limed hot-rolled bar through a die by means of a so-called draw bench mechanism. This bench may be of any convenient length, depending upon the cold-drawn lengths desired but usually is from 30 to 75 feet long. This mechanism consists of three main parts, namely, the drive, the pointing head and the pulling head. The drive involves a variable speed electric motor, driving a series of gears through a sprocket which in turn operates an endless chain running from one end of the bench to the other passing over an idle sprocket on the extreme end. The pointing head is a mechanism which grips the bars to be drawn and pushes them through the dies. It consists of wedge-like grips which are so arranged that resistance of the dies tightens the grip on the bars and makes the action more positive. The pointing head is drawn forward by side rods, actuated by cams on the idle sprocket shaft which is connected to the main drive. As the head is moved forward, the bars are pushed through the dies sufficiently to permit gripping by the draw head.

The jaws of the draw head carriage then grip the end of the bar and the carrier is hooked to the endless block chain which furnishes the drawing power for the cold drawing operation itself.

In addition to the "pushing" method of introducing the bar into or through the die, pointing by cold swaging or by "over-pickling" is also employed. The pointing by swaging is usually handled on a rotary type pointing machine effecting about 1/16-inch reduction of

diameter for a distance of 6 to 8 inches back from the bar end. Pointing by "over-pickling" is employed usually on special sections which cannot be swaged and will not permit pushing. This is done by standing the bars on end to a depth of 6 or 8 inches in small pickle boxes in hot sulphuric acid solutions as high as 50 per cent concentration at the start of action. The acid concentration weakens as action progresses without acid additions so that at the completion of the operation the solution may be as weak as 5 to 10 per cent. By this means, it is possible to reduce the cross section about 1/8 inch for a distance of 6 to 8 inches, sufficient to permit passing the bar freely through the die for gripping by the draw head.

In addition to the lime coating present as a result of the liming operation, it is necessary to lubricate the bar further in cold drawing. Various methods and lubricants are used although the general practice is to use either grease, held in a box directly in front of the die through which the bar passes, or if the grease box is not used, a stream of soluble oil solution is directed against the bar just prior to entry in the die. A good grade of engine oil or cottonseed oil is also employed at times for special purposes, although this method is more expensive. The lubrication employed together with the lime coating not only promote better cold-drawn finish but also ease die wear, thereby contributing to longer die life.

The speed of drawing varies widely, depending upon drawing equipment, bar size and section, quality of finish desired, chemical grade, physical limitations of the metal being handled, amount of reduction or draft employed, etc. A fairly representative range for commercial drawing speeds would be from 40 to 100 linear feet per minute, although for some sections such as flats, the speed might be as low as 9 feet per minute and in wire sizes as high as 300 feet. One to four bars are drawn at a time, depending upon cold drawing facilities and the size and character of the stock being handled.

The usual draft or reduction of diameter employed in cold-drawing varies from 1/32 to 1/8 inch from hot-rolled to cold-drawn. For the most part, the range used is 1/32 to 1/16 inch. Once again the amount of draft employed depends upon such variables as bar size, section, chemical grade, physical characteristics and finish required.

Dies for Cold Drawing—While it is not the intention to discuss dies, nevertheless in passing it might be well to call attention to several general features regarding this important feature. Design, chem-

ical composition of the die material, heat treatment and fabrication methods vary considerably. However, it might be stated that the dies being used commercially today are either of the solid or set-up type. A sketch of these types will be found in Fig. 1 which also gives some idea of the drawing action. Rounds are drawn in the solid type and also hexagons and squares in sizes up to approximately 1¼ inch. The set-up type is used for the larger sizes of hexagons and squares and flats.

Straightening—While much can be accomplished by the operator in proper alignment of the work to prevent undue warpage resulting

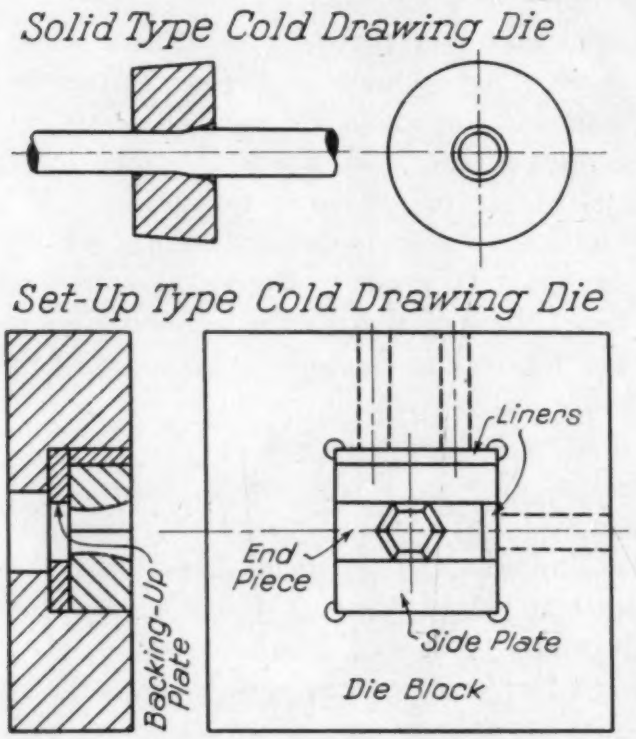

Fig. 1—Sketch Showing Types of Drawing Dies.

from the strains of cold drawing, it is imperative that the cold-drawn product be straightened to meet the exacting straightness requirements demanded.

For cold-drawn rounds, by far the most popular type of straightener is the Medart. Briefly, this machine is a cross roll type consisting of one concave and one straight roll. The bar to be straightened is fed horizontally between the cross rolls which if set to give sufficient pressure to straighten, propel it through the rolls at a feed per bar revolution dependent upon the bar size and angle of the rolls.

Not only is the bar straightened to fairly close accuracy but it is also polished to some extent due to the burnishing or wiping action of the rolls.

Smaller sizes of rounds, approaching wire sizes, are straightened, usually by means of straight rolling between a series of top and bottom rolls arranged horizontally but staggered.

This type of straightening does not polish the bar and in cases where the polishing effect is desired even the smaller sizes are handled in the Medart type. Where extremely close straightness limits are to be obtained, bars may be straightened by hand or power gagging although this type of straightening is most popular for the larger sizes of turned and polished or turned and ground rounds.

Hexagons, squares, flats and special shapes are straightened either by means of the staggered roll type straightener or by gagging, depending upon the section and the requirements to be met.

Cutting—The bar lengths handled throughout the cold drawing procedure usually are figured to give several multiples of the length ordered by the buyer. For example, if 12-foot lengths of 1-inch round were ordered, the hot-rolled bar lengths would be fixed to give two or three multiples of 12 feet each cold-drawn. Shearing, cracker-shearing, sawing and lathe cutting are the popular methods of cutting. It is true that in shearing the section is distorted slightly at the shear cut and it is almost impossible to avoid a slight kink near the end of a sheared bar, but the severity of these features is not of sufficient magnitude to be objectionable in most cases. However, where absolute freedom from these features is desired or where the bar sizes to be cut are large enough to cause undue distortion, either sawing or lathe cutting is employed.

Slushing—Obviously, the highly finished cold-drawn bar product, one of whose requirements is a bright finish, must be protected against rusting. As a result of the lubricant used in drawing and light oils used to protect the Medart rolls in the case of Medart straightening, practically all bars bear a thin protective coating of oil or grease. However, this film is not uniformly distributed, likely has been rubbed off completely in spots as a result of handling and in no case offers sufficient protection against rusting. Consequently, all cold-drawn material is slushed with a protective coating of oil before shipment unless otherwise specified. The three popular slushing methods are spraying, dipping and brushing.

The desirable slushing fluid is one which gives a thin uniform

film capable of withstanding atmospheric temperature changes without cracking or running thin in spots. Either condition would leave the bar exposed, and rusting would set in. There are many formulae covering slushing fluids designed to give best results and so varied are the ideas in this connection that no attempt will be made to make definite recommendations. In general, in spite of the great susceptibility of cold-drawn bars to rusting and the unfavorable conditions often encountered in shipping, subsequent handling and stocking, slushing has been developed to the point where the danger of rusting can hardly be considered a major hazard.

Inspection—The quality requirements pertaining to cold-drawn products today are very exacting and the matter of meeting specifications demands costly and extensive inspection practices. Such inspection protection extends back as far as the iron-making and is maintained through steel-making, blooming, billet reconditioning, hot-rolling and cold finishing. Every practical inspection precaution is taken to guard against the shipment of material apt to draw unfavorable criticism from the purchaser, regardless of the peculiarities of his requirements.

Effects of Cold Drawing and Cold-Drawn Characteristics

Finish—The cold-drawn bar, due to the nature of its processing, is free of scale, smooth and bright. As a general rule rounds, because of the polishing action of the Medart, possess a brighter finish than other sections which do not receive the benefit of such work. However, demands for high-class finish have developed practices which give a very satisfactory finish reasonably free of objectionable pitting, surface defects and any visible irregularities inclined to mar the appearance of the bar. These characteristics make the product desirable for shafting and machined parts, particularly if the finished parts retain a portion of the original bar surface. Furthermore, freedom from scale increases cutting tool life and precludes the possibility of loose scale contaminating cutting lubricants with danger of subsequent injury to bearings and moving machine parts.

Size-Accuracy—In Table II is shown the standard manufacturing tolerances covering cold-finished steel bars and shafting. A study of this table will show the much greater degree of accuracy obtained in cold drawing as compared with hot rolling. In Table III is shown a comparison of the standard permissible size variations of several sizes of hot-rolled and cold-drawn rounds. So pronounced is this dif-

Table II
Standard Manufacturing Tolerances

Cold-Finished Steel Bars And Shafting
(0.50% Carbon And Less)

Rounds	0.30% Carbon Or Less	0.31% To 0.50% Carbon
1 inch or under	0.002 inch minus	0.003 inch minus
1 1/16 to 2 inch inc.	0.003 inch minus	0.004 inch minus
2 1/16 to 4 inch inc.	0.004 inch minus	0.005 inch minus
4 1/16 to 6 inch inc.	0.005 inch minus	0.006 inch minus
6 1/16 to 7 inch inc.	0.006 inch minus	0.008 inch minus

Standard Manufacturing Tolerances On "HEAT TREATED" Bars Are Double Those Shown

Hexagons	0.30 per cent Carbon Or Less	0.31 per cent To 0.50 per cent Carbon
Up to 0.3 inch inc.	0.002 inch minus	0.003 inch minus
Over 0.3 inch to 1 inch inc.	0.003 inch minus	0.004 inch minus
Over 1 inch to 2½ inch inc.	0.004 inch minus	0.005 inch minus
Over 2½ inch to 4 inch inc.	0.005 inch minus	0.006 inch minus

Squares	0.30 per cent Carbon Or Less	0.31 per cent To 0.50 per cent Carbon
Up to 0.3 inch inc.	0.003 inch minus	0.004 inch minus
Over 0.3 inch to 1 inch inc.	0.004 inch minus	0.005 inch minus
Over 1 inch to 2½ inch inc.	0.005 inch minus	0.006 inch minus
Over 2½ inch to 4 inch inc.	0.006 inch minus	0.008 inch minus

Flats	Narrower Than 1 ½"		1 ½" To 4" Wide		Over 4" To 6" Wide	
Thickness	0.30 per cent Carbon Or Less	0.31 per cent To 0.50 per cent Carbon	0.30 per cent Carbon Or Less	0.31 per cent To 0.50 per cent Carbon	0.30 per cent Carbon Or Less	0.31 per cent To 0.50 per cent Carbon
Up to 0.3 inch inc.	0.003 inch minus	0.004 inch minus	0.005 inch minus	0.006 inch minus	0.008 inch minus	0.010 inch minus
Over 0.3 inch to 1 inch inc.	0.004 inch minus	0.005 inch minus	0.005 inch minus	0.006 inch minus	0.009 inch minus	0.012 inch minus
Over 1 inch to 2½ inch inc.	0.005 inch minus	0.006 inch minus	0.006 inch minus	0.008 inch minus	0.010 inch minus	0.012 inch minus

The tolerances on flats apply to thickness as well as width, and the width dimension determines the tolerance

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Table III
Size Accuracy of Cold-Drawn Versus Hot-Rolled

Size	Variation	Tolerance	Out of Round
½" Rd. H.R.	0.492 inch to 0.508 inch	± 0.008 inch	0.012 inch
½" Rd. C.D.	0.500 inch to 0.498 inch	0.002 inch minus	0.001 inch
1" Rd. H.R.	0.990 inch to 1.010 inch	± 0.010 inch	0.015 inch
1" Rd. C.D.	1.000 inch to 0.998 inch	0.002 inch minus	0.001 inch
2" Rd. H.R.	1.978 inch to 2.022 inch	± 0.022 inch	0.030 inch
2" Rd. C.D.	2.000 inch to 1.997 inch	0.003 inch minus	0.0015 inch
3" Rd. H.R.	3.000 inch to 3 5/64 inch	5/64 inch over	3/64 inch
3" Rd. C.D.	3.000 inch to 2.996 inch	0.004 inch minus	0.002 inch

Differences of even greater proportions exist between hot-rolled and cold-drawn hexagons, squares, flats, and special sections.

ference that little further attention need be paid this feature, other than to state that the close accuracy of the cold-drawn product is one of its important advantages.

Straightness—While no standard table of straightness limits is published, the degree of straightness required by cold-finished steel users is very close and while for the most part straightness requirements vary considerably, depending upon the peculiarities of the application, no serious difficulty is encountered in obtaining very close accuracy in this connection.

Hot-Rolled versus Cold-Drawn Physical Properties

Cold drawing changes the physical characteristics of steel profoundly. The tensile and yield values are raised, the increase in yield being proportionately greater than the increase in tensile strength. Fig. 2 shows a typical stress-strain diagram covering three popular chemical grades and a general idea of the effect of cold drawing on

Table IV
Representative Physical Properties
Hot-Rolled Versus Cold-Drawn

Grade	Size	Yield Point	Tensile Strength	Elong. Per Cent	Red. Per Cent	Brinell
SAE 1112	1 1/16" Rd. H.R.	40480	64780	35.0	53.0	131
SAE 1112	1" Rd. C.D.	82460	89700	17.0	43.5	187
SAE 1020	1 1/16" Rd. H.R.	38240	59100	37.0	62.0	118
SAE 1020	1" Rd. C.D.	75120	82750	20.0	51.7	156
SAE 1035	1 1/16" Rd. H.R.	45240	77100	35.0	48.0	149
SAE 1035	1" Rd. C.D.	87240	94240	15.2	41.2	197

The above cold-drawn values obtained on hot-rolled bars cold-drawn at 1/16 inch draft amounting to 11.4 per cent in regular commercial production.

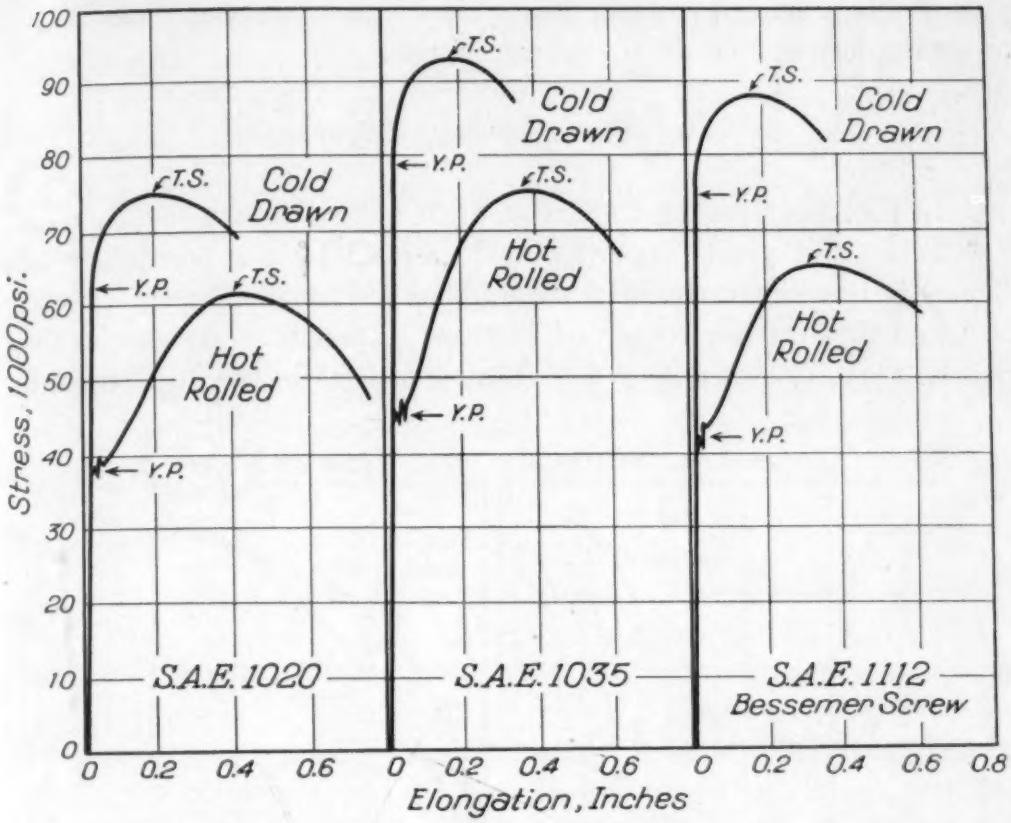

Fig. 2—Typical Stress-Strain Diagrams of Hot-Rolled and Cold-Drawn Steels.

tensile and yield properties may be obtained therefrom. In general, since hardness and fatigue properties are proportional to tensile strength, these features are likewise affected by cold drawing. The specimens used to obtain the results in Fig. 2 were 1-1/16 inch rounds hot-rolled, cold-drawn to 1-inch round in regular commercial production.

As might be expected the increase in tensile, yield and hardness values resulting from cold drawing are attended by a decrease in elongation and reduction values. In Table IV will be found representative physical properties of three popular hot-rolled and cold-drawn chemical grades. The characteristic changes in physical properties mentioned above will be noted. A closer study of Table IV and Fig. 2 will indicate that the Bessemer screw grade has undergone more drastic changes than the two open-hearth grades listed. While there are definite differences in sensitivity to cold work between heats of the same grade whether in open-hearth or Bessemer, in general, the Bessemer steels are appreciably more sensitive under cold work than the open-hearth steels. Deliberate control of the sensitivity

property is practiced more or less widely where considered necessary to obtain desired cold-drawn characteristics.

Effect of Draft on Physical Properties

In Fig. 3 is shown the effect of draft or cold-drawn reduction on the tensile and yield properties as measured by the percentage of change in these characteristics from hot-rolled to cold-drawn. It will be noted that the percentage of increase in tensile is greatest in the S.A.E. 1112, followed by S.A.E. 1020 and 1035 in the order named.

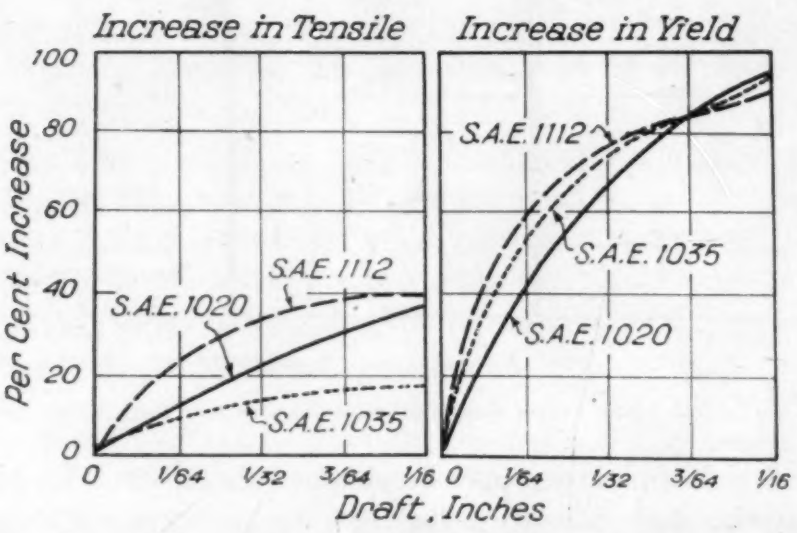

Fig. 3—Effect of Draft on Tensile Strength and Yield Point.

It will also be noted that the greater sensitivity to cold work of the S.A.E. 1112 is indicated by the sharper increase of tensile strength at the lighter drafts. The S.A.E. 1112 behaved similarly in its change of yield value. The S.A.E. 1035, however, showed proportionately a greater change in yield than tensile as compared with the S.A.E. 1020 which is representative of the normal behavior of these grades. The curves shown are meant to give an idea of the general behavior of these grades under commercial practices. The use of special practices and steel making, hot rolling, normalizing or annealing before cold drawing affect the sensitivity feature appreciably. Obviously, deliberate control of the processing practices is used for specific influence on the cold-drawn characteristics.

In general, greater drafts than those covered by Fig. 3 would not result in higher tensile values on the average but the yields would

continue to increase rather sharply until breakage of the bar in drawing occurred.

Penetration of Cold-Drawing Effect

There are those who maintain that the change of physical characteristics effected by cold drawing is only skin deep, or at best ex-

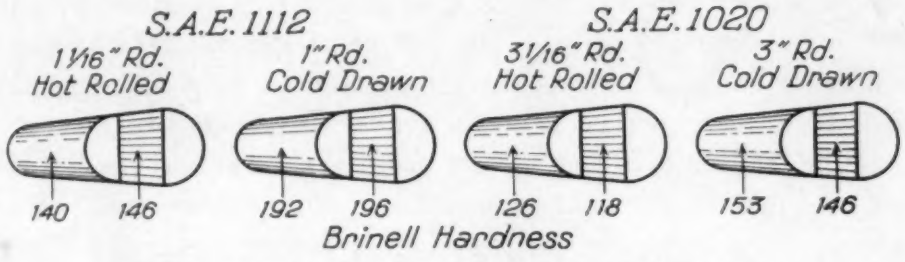

Fig. 4—Hardness Penetration Effected by Cold Drawing.

tends only a short distance under the bar surface. Such an idea is entirely erroneous. In general, it might be said that the effect of cold drawing is uniformly distributed over the cross section even in the largest diameters cold-drawn. In Fig. 4 will be found sketches of two sizes of hot-rolled and cold-drawn cross sections showing the Brinell results obtained on the surface and at the center of the pieces. These results, as will be noted, indicate the uniform changes of physical characteristics effected in cold drawing.

Effect on Machinability—It is generally agreed that cold drawing has a beneficial influence on inherent machining quality. However, aside from inherent machining quality, the cold-drawn bar product with its close size and section accuracy, freedom from scale and better straightness offers sufficient machining advantage over hot-rolled material, particularly in automatic screw machine work to make it preferable and in most cases more economical than the lower priced hot-rolled stock where ease of machining is a factor. In support of this statement is the steady demand for cold-drawn material for machining work in the face of serious attempts to accommodate hot-rolled material in the interest of lowering raw material costs.

General

Up to 0.50 per cent carbon, all of the widely used plain-carbon grades are cold-drawn commercially. The most popular grades are

S.A.E. steels 1112, X-1112, 1020, X-1020, 1035, 1040, 1045, 1120, X-1314, X-1315, 1335 and X-1335. Many users do not identify their steel grades by S.A.E. designation. However, regardless of the method of identification, the grades listed, allowing for slight chemical preferences, are representative of the most popular commercial grades. There are those who suspect that the hot-rolled stock is annealed or normalized preparatory to cold drawing. As a matter of regular procedure, such is not the case although if specified by a purchaser or necessary because of the nature of the required properties, annealing or normalizing, as the circumstances may require, is employed. However, in carbon steels over 0.50 per cent carbon, it has been found desirable, in fact almost necessary, to anneal the hot stock before cold-drawing, primarily to soften it so as to minimize breakage in cold drawing. Further in these higher carbon steels, the ductility of the cold-drawn product is considerably more desirable in the annealed and cold-drawn state than if cold-drawn from unannealed hot-rolled stock.

Within the last few years, some advantage is realized in special instances by low temperature treatments after cold-drawing. Such treatments may involve temperatures ranging from 300 to 1000 degrees Fahr. or 1100 degrees Fahr., depending upon the results sought. In some instances "strain drawing" at temperatures in the neighborhood of 400 to 700 degrees Fahr. is employed as an "accelerated aging" function. On the other hand "tempering" effects with improved ductility without appreciable losses in tensile and yield values are produced through the use of temperatures between 700 and 1000 degrees Fahr. However, it should be understood that heat treatment, either before or after cold drawing, is not regular procedure, but could be classified, except in the case of higher carbons, as special procedure to meet special requirements.

There are those who believe the metallographist can identify cold-drawn material from hot-rolled with the microscope through identification of elongated grains in a longitudinal section. As a general thing, this is not so. Even in sizes as small as ½ inch in diameter, the amount of draft employed is not sufficient to cause enough structure distortion to be apparent under ordinary microscopic magnification. It is true that in carbon steels over 0.50 per cent carbon under some conditions slight indications of grain elongation may be detected to a shallow depth but as stated before, in general such grain distortion as might have been effected in cold drawing is so small as to be

practically negligible. This is not true in some wire drawing where the drastic cold-reductions employed do leave their mark on the wire structure.

There are those who suspect that the skin of the bar is removed in some manner when drawn through the die. Obviously this is not true. No metal is removed, other than the scale in pickling, during the entire process. The hot-rolled bar is actually stretched through the die. For example, a 1-1/16-inch round 33 feet long will elongate about 3 feet when cold-drawn to a 1-inch round, thereby giving a cold-drawn length of slightly over 36 feet.

Conclusion

The uses of cold-drawn steels have become so wide that their applications are too numerous to attempt to cover completely. Some of the uses are for spark plug shells, radio speaker cores, bolts, nuts, screws, gears, pinions, typewriter, sewing machine and cash register parts, chain pins and rolls for silent chains, motor and transmission shafts, motorcycle crank axles, oil pump shafts and gears, piston pins, steering gear arms, bushings, pins and worms, tie rods, wringer rolls, gun parts, etc. Industries manufacturing automobiles, motorcycles, bicycles, sewing machines, typewriters, adding and calculating machines, textile machinery, shoe machinery, electric refrigerators, railroad equipment, agricultural implements and electrical equipment are large consumers.

A better and more complete appreciation of the characteristics of cold-drawn material should enlarge its field of application. It is the aim of the steel manufacturer, through exhaustive studies of manufacturing practices, to develop definite processing procedure to produce specific properties. Considerable progress has already been made in this connection and with the passing of time it is only reasonable to expect a more complete and precise understanding which will prove mutually advantageous to both consumer and producer.

COLD DRAWING SEAMLESS STEEL TUBES

By Horace C. Knerr

Abstract

In this paper the author has attempted to give only a brief outline of elementary principles involved in the art of producing cold drawn seamless steel tubing, touching upon methods, equipment, and tools and giving a representative example of the flow of metal, grain size, and physical properties resulting from cold drawing. A few of the more important avenues for research and development are suggested.

THE art of seamless tube drawing as it exists today is largely the result of practical evolution during half a century, served by great ingenuity on the part of shop men, but without the aid of a proportionate amount of scientific research. Remarkable results as to form, accuracy and finish have been achieved, and many improvements made in method, but metallurgical and mechanical problems are numerous, complex, and exacting and offer a fertile field for improvement in production and product and reduction of costs through technical study.

Metal tubing owes its great and ever widening usefulness principally to three important characteristics:

1. Efficiency as a conduit for fluids.
2. Maximum strength in bursting, torsion, bending, and compression (imparting also the quality of lightness as a structural material).
3. Mechanical accuracy.

Seamless tubing is closely seconded or even surpassed in some applications by high grade welded tubing, the latter usually being cold drawn one or more times for finish or size.

Sizes of cold-drawn tubing range from the neighborhood of 10 inches diameter 1 inch wall down to .01 inch diameter .001 inch wall—a ratio of 1,000 to 1 in diameter or wall thickness and 1,000,000 to 1 in sectional area.

A paper presented as part of the Symposium on the Plastic Working of Metals, Eighteenth Annual Convention of the Society held in Cleveland, October 19 to 23, 1936. The author, Horace C. Knerr, is consulting metallurgical engineer, and President of Metlab Co., Philadelphia. Manuscript received July 15, 1936.

The first cold-drawn seamless tubing produced in this country was made by the late Mr. Ellwood Ivins, then a young Englishman of mechanical genius, about 1887, for George Westinghouse, founder of Westinghouse Electric and Manufacturing Company. Mr. Ivins also founded a company bearing his name, in Philadelphia, the oldest tube mill in America, which now operates as a specialty mill with a product of great diversity and refinement.

The author is indebted to Mr. John B. Cording, President, and Mr. Horace S. Kircher, Metallurgist, of Ellwood Ivins Steel Tube Works for generous assistance in the preparation of this article.

General

The manufacture of seamless tubing grew out of the very old art of wire drawing, plus the technique of producing a hollow bar, billet, or crude tube to start with. Early methods included drilling from bar, cold cupping from sheet, casting a hollow ingot, and hot punching, which are still used in special cases. In quantity production they have been replaced by various ways of hot piercing, principally the Mannesman process.

As in cold drawing wire or rod, the previously pointed end of the crude tube is put through the opening of a round, tapered die whose diameter is somewhat smaller than the (present) outside diameter of the tube. The point is grasped in a pair of jaws, tongs, or pliers and the tube is pulled through the die, causing a reduction in diameter and cross sectional area and an increase in length. This operation is repeated until the desired final outside diameter is reached. A mandrel may or may not be used to control the inside diameter.

Between successive passes, the tube is usually annealed to resoften the metal, pickled to remove annealing scale, baked to dry and remove hydrogen embrittlement, and lubricated. This cycle may be repeated as many as twenty times in producing very small tubes (for example in reducing a hot-rolled tube of 2-inch O.D., #11 BWG [.120 inch] wall to the hypodermic needle size, .025 inch O.D., .005 inch wall.)

Process

Cold-drawn tubing is usually manufactured to order, under an immense variety of specifications for size, wall thickness, dimen-

sional tolerances, and shape, not to mention composition and physical properties.

Pointing

The raw material of the cold-drawn tube mill is the hot-finished tube of random mill length, referred to as a "billet" and usually stocked in a moderate variety of sizes. Having selected from stock for moderate production, or ordered from the hot mill for large production, billets of diameter and wall thickness suitable for reduction to the desired finished product, they are cut into shorter lengths and pointed. Pointing or "tagging" is usually done hot in various ways, including vertical hammers and toggle presses for the larger sizes, and by swaging between rotating hammer dies for the smaller sizes. When successive passes through the die have elongated the tube beyond a convenient length, it is cut in two and the new end pointed. The original point must sometimes be forged down smaller to enter the successively smaller dies.

A small hole may be punched near the tag or tang end during the pointing operation to facilitate access of pickling fluid and later of lubricant to the interior.

Pickling

Scale from hot rolling, pointing, or annealing is removed by pickling in dilute (5 per cent) sulphuric acid, followed by rinsing and baking to dry and remove hydrogen embrittlement. The tube may be dipped in lime emulsion to neutralize acid and contribute to lubrication.

One of the most important practical features of the pickling operation, often neglected, is the complete removal of loose as well as fixed scale from the inside and outside of the tube. Failure to do this causes rapid wear and break down of dies and mandrels, scratched walls, siezing and breakage of tubes, and other difficulties. Dirt or scale mixed with the lubricant or adhering to the lubricated tube is also a common source of trouble.

Lubrication

The tube is lubricated or "doped" by immersion in a vat containing a mixture such as palm oil and tallow, or directly at the draw bench by means of oil and a circulating pump. The character of lubrication is extremely important.

Certain metallic lubricants for coating metal are very efficacious if they are alloyed with the tube so as not to strip off. For example, copper is applied by dipping in a copper sulphate solution after pickling, tin by electro plating, and lead by a chemical flux and hot dip. The cost of application and final removal of these coatings is important and detracts from their benefits. A lightly oxidized coating is, in some instances, an aid to lubrication. Much has been done, but lubrication, metallic and nonmetallic, remains a fertile field for research.

Annealing

In most cases the tube must be annealed after each cold pass. Occasionally two passes may be made without annealing, and under special conditions even more.

Annealing consists fundamentally in causing grain growth or readjustment of the ferrite, since the carbide particles are too hard to be elongated by cold drawing, although they may be distributed lengthwise. Annealing may therefore be done at moderate temperatures, and without very slow cooling. In practice it is done in open furnaces, in retorts and in furnaces with atmosphere control affording greater or less degrees of surface brightness, according to circumstances. A first anneal of the hot-worked tube above the critical range with slow cooling, to spheroidize carbide, is desirable.

Pickling after annealing is the principal "bugbear" of the cold drawing tube mill. Strenuous efforts are being directed toward its elimination by securing a bright anneal. Where bright annealing is done the lubricant must sometimes be modified to secure adherence to the smooth, unpickled, surface.

Methods

The inside diameter of the die, of course, determines the outside diameter of the tube. The method of handling the inside diameter of the tube divides tube drawing into three classes:

1. Without a mandrel.
2. With a fixed (short) mandrel or plug.
3. With a moving (long) mandrel or rod.

Sinking

The first method is known as "sinking" and is used where the

inside diameter need not be closely controlled, where smooth inside finish is not required, where a thickening of the wall may be called for or where the wall is very thick, relative to diameter, as in capillary or high pressure tubing. Inside lubrication is eliminated. A reduction of 20 to 35 per cent in cross sectional area of the tube per pass is customary.[1] For example:

Before pass: 1½ inch O.D., #12 BWG (.109 inch) (1.62 pounds per foot).

After pass: $1\frac{1}{16}$ inch O.D., #12 BWG (.109 inch) (1.11 pounds per foot).

Reduction: 31.5 per cent.

In light gages, such as #20 BWG (.035 inch) or less, it is not practical to reduce more than about 20 per cent per pass, by sinking.

Too severe sinking causes roughening (longitudinal grooving) of the inner surface and irregularities of wall thickness.

Short Mandrel or Plug

The second method is perhaps the most general and representative. A short mandrel, attached to a slender rod, is held so that its working end is in a plane with the die. The back end of the rod is adjustably held in a bracket at the back end of the draw bench. By moving the mandrel back and to one side, the tube may be slipped over it. The tube point is then inserted in the die and cold drawing proceeds. The mandrel determines the inside diameter of the tube. In this process, both the outside and inside diameter must be reduced in each pass. The wall thickness may or may not be reduced. Friction between the tube walls, outer and inner, and the die and mandrel is very great, and becomes a determining factor.

A thin tube has nearly double the friction surface of a solid bar of equal diameter and only a small fraction of its sectional area or pulling strength. The material in the tube wall is cold-worked to a fairly uniform degree throughout its thickness. Therefore the amount of possible reduction in one pass and the number of passes between successive annealing operations, both primary factors in production cost, are to a great extent determined by surface friction.

This method cannot be used for tubes of very small diameter (less than about ¼ inch) partly because of the difficulty of attaching the mandrel to its retaining rod, and because of the greater surface

[1]Note: Examples are taken from tube mill practice. Reductions are figured on basis of weight per foot from tube data tables.

area of the tube wall in proportion to the sectional area of the rod. Customary reduction of area per pass is 30 to 40 per cent.

A typical operation would be:

Before pass: 1½ inch O.D., #12 Ga. (.109 inch) (1.62 pounds per foot).

After pass: 1⅜ inch O.D., #15 Ga. (.072 inch) (1.00 pounds per foot).

Reduction: 38 per cent.

Or another:

Before pass: ⅞ inch O.D., #16 BWG (.065 inch) (.56 pounds per foot).

After pass: $\frac{13}{16}$ inch O.D., #19 BWG (.042 inch) (.35 pounds per foot).

Reduction: 37½ per cent.

Long Mandrel or Rod

In the third method the mandrel consists of a long hard rod or wire of uniform diameter which is inserted into the inside of the tube before drawing, but is not held at the back end. It travels forward with the tube as the latter is drawn down hard upon it in passing through the die, making the tube and mandrel like a solid piece. Frictional drag is much less than with the short stationary mandrel. One or more extra operations are required, because the mandrel must be removed from the tube after drawing. This is done by reverse rolling through grooved rolls, spiral rolling through straightening rolls, or passing through rotating swaging dies, etc., so that the tube will be slightly enlarged to free it from the mandrel and permit slipping off.

The long mandrel moves forward with the front or reduced portion of the tube, and therefore travels faster than the rear, undrawn portion. This presents a continually fresh surface to the interior of the tube where it enters the die, reduces local heating, and promotes lubrication. Wear on the mandrel surface is not nearly so great as on the short mandrel, therefore the material need not be so hard. Cold-drawn piano wire is used for the smaller sizes down to hypodermic needles, and specially hardened and ground alloy steel rod for the larger, from $\frac{3}{16}$ inch up to a maximum of about 1-inch diameter. Customary reduction by this method is 40 to 50 per cent. Example:

Before pass: 7/8 inch O.D., #16 BWG (.065 inch) (.56 pounds per foot).

After pass: 13/16 inch O.D., #20 BWG (.035 inch) (.29 pounds per foot).

Reduction: 48 per cent.

In addition to the smaller sizes where a fixed mandrel is not practicable, this method is used for intermediate sizes where the wall is very thin relative to the diameter (where the excessive frictional pull in the fixed plug would injure the tube wall) such as in tubes for the manufacture of corrugated flexible tubing, Bourdon gages, golf shafts, etc.

The difficulty of obtaining large rods of satisfactory hardness and straightness, and their cumbersome weight has hindered the use of this efficient method in the large sizes of tubing. Progress is being made in overcoming this limitation.

Draw Bench

The draw bench as used for medium and large size tubing is illustrated diagrammatically in Fig. 1,[2] and in the photographs, Figs. 2 and 3. Various modifications of this essential principle exist. The

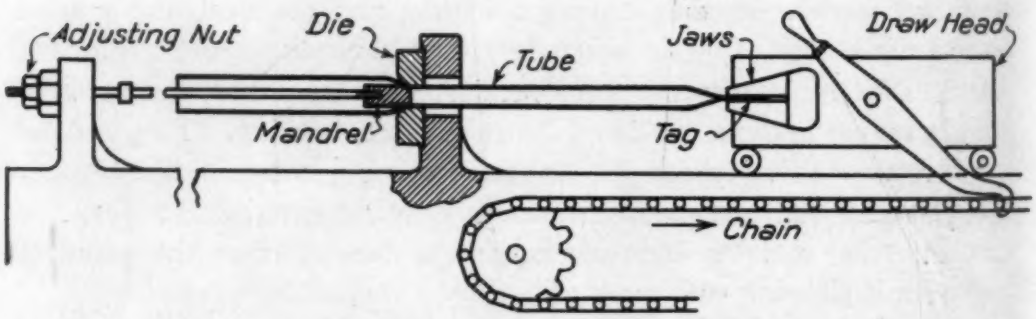

Fig. 1—Diagrammatic View of Draw Bench Showing Seamless Tube in the Process of Drawing.

base of the bench is of cast iron or structural steel, perhaps 40 to 60 feet long. A vertical bracket, frame, or "die head" at the center supports the die, with provision for easily changing dies. Jaws for clamping the point are carried in a draw head which moves back and forth along tracks on the front bench, actuated by a continuous, motor driven chain, to which it is engaged by a hook. Means are provided for automatically disengaging the hook when the tube has

[2]Figures 1, 2, 3, and 4 are reproduced from an article "Story of Aircraft Tubing," S. L. Gabel and H. C. Knerr, *Aviation*, March 29, April 12, April 26, 1930.

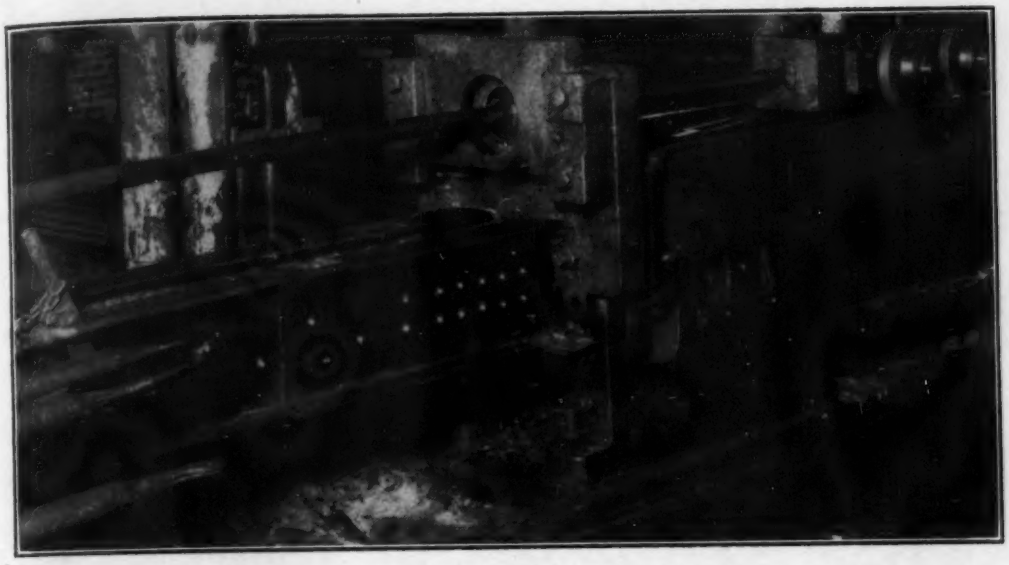

Fig. 2—Center Part of Draw Bench Showing Tube Partly Drawn.

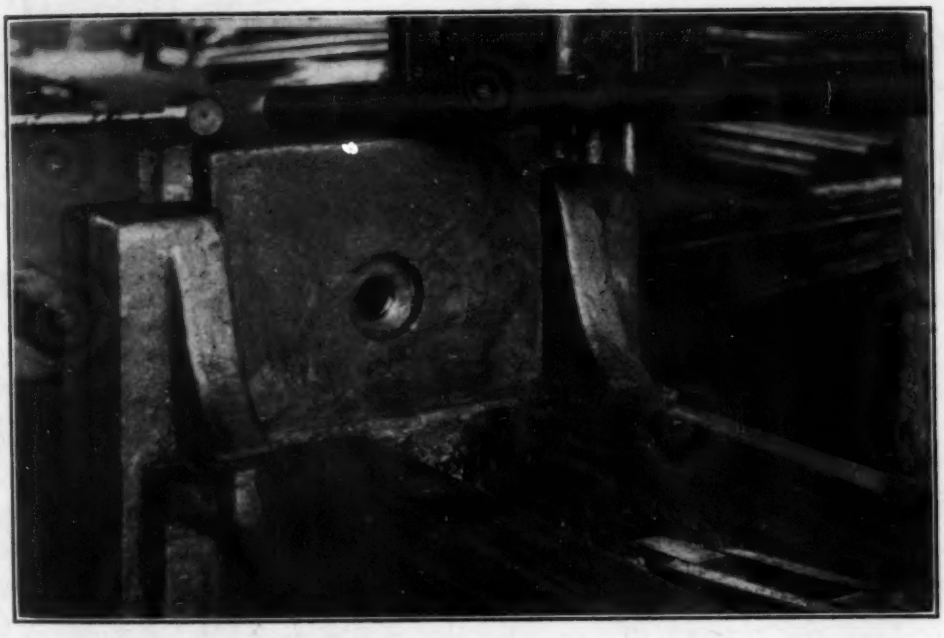

Fig. 3—Tube Ready to Slip Over Short Mandrel.

been pulled through the die. The draw head is returned to position in front of the die by hand or auxiliary mechanism. Benches are sometimes made double so that two tubes may be pulled at the same time.

A hydraulic piston may be substituted for the chain and draw head and is quieter, faster, and better adapted to automatic operation.

A bracket on the back end of the bench holds the rear end of the

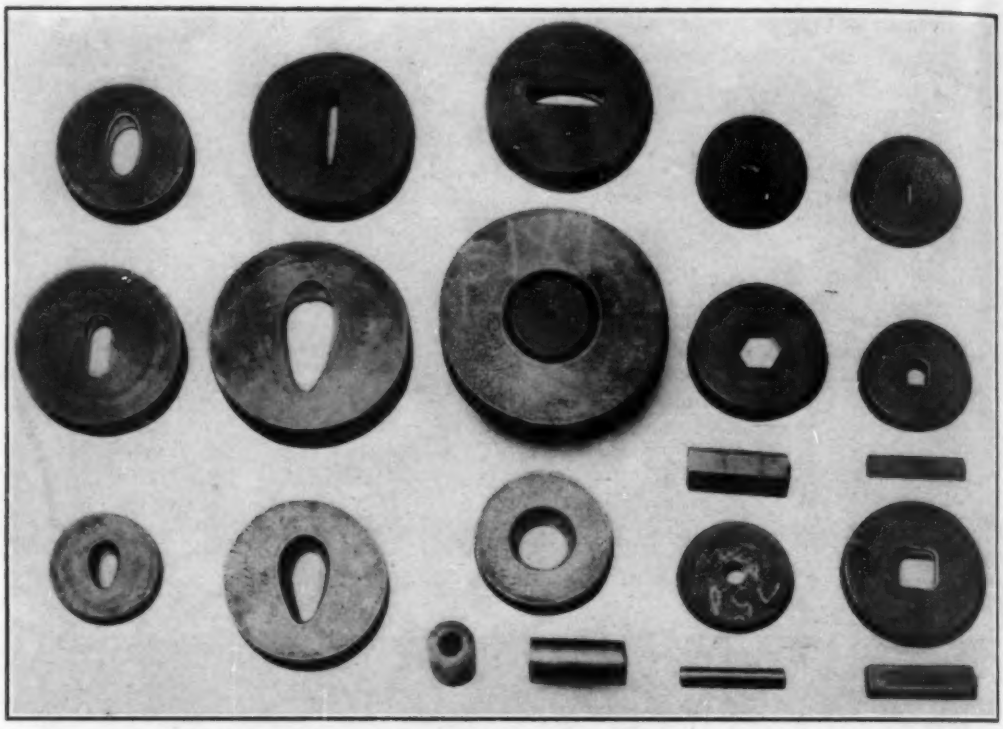

Fig. 4—Typical Dies and Mandrels.

rod on whose forward end the short mandrel is secured. An adjusting nut permits bringing the mandrel in correct longitudinal position relative to the die. The tube itself centers the mandrel. A stop near the back end of the rod permits moving the mandrel back far enough to slip the tube on, but prevents it from flying back too far by elastic recoil when the tube leaves the die.

The same or a similar bench is used for sinking, or for drawing over a long mandrel, the rear bracket being omitted.

Speed of travel of the draw head varies widely according to the size of the tube and the character of the material and lubricant. Ten feet to sixty feet per minute would cover the usual ranges.

Tubes of very small size are coiled and drawn like wire, on reels, instead of straight draw benches.

Die Forms

Special shapes are produced by drawing a round tube through a forming die, usually, but not always, without a mandrel. The correct proportioning of die and choice of diameter and wall thickness require skill and experience. Dies and mandrels for a few of many shapes produced are illustrated in Fig. 4.

Die Contours

The design of die contours seems to have been largely empirical, evolving from wire drawing practice. Until rather recently, dies were made from 0.90 to 1.10 carbon steel upset forgings, bored and turned in the tool room, more or less according to the ideas of the tool maker or shop foreman. They were hardened by jet quenching, and repolished with emery. After drawing a few hundred feet of tubing, wear caused enlargement beyond tubing tolerance and they were "shrunk" by re-quenching. This repeated "upsetting" of the

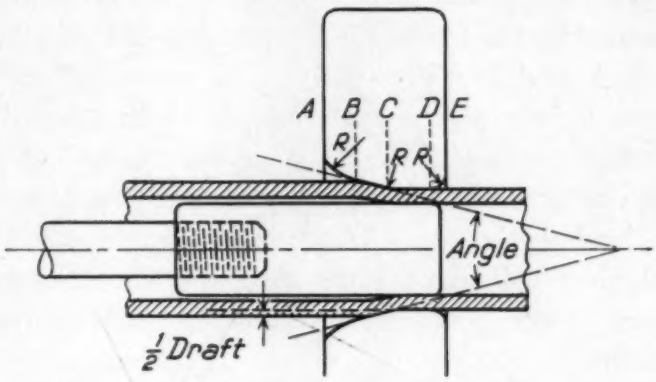

Fig. 5—Diagram Showing Elements of Die Contour. Draft—Previous size minus die size. Angle—Determines metal flow. AB—Bell guides point and distributes lubricant. BC—Breakdown bearing or area of reduction. CD—Pull bearing, die diameter. DE—Relief which allows for slight misalignment.

inner surface due to volumetric displacement during the quench, caused deformation of die contour and longitudinal cracking of bore. The dies were then annealed and re-bored to a new size.

Special alloys are being adopted for the larger sizes (for example carbon 1.30 per cent, tungsten 3.5 per cent) and tungsten carbide for the smaller sizes. Nitrided tool steels of certain analyses have given excellent service.[3] The high cost of tungsten carbide dies, and the fact that their contour, once established, is not readily altered, has stimulated some study of contour design.[4]

The elements of die contour are illustrated in Fig. 5. The factors—draft, angle, break down bearing length, and pull bearing length for tubes of various diameters, wall thickness and material—are determined according to empirical formulae, or the ideas of the individual mill.

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[3]"Nitrided Tool Steels," H. C. Knerr, *Iron Age*, Sept. 26, 1936.

[4]See Booklet, Union Wire Die Corporation.

The diversity of factors renders the problem of best die contour quite complex. There is evidently a fertile field for research and analysis here, based upon the flow properties of the metals drawn.

Metal Flow

For observation of metal flow and microstructure, low carbon commercial ("Shelby") tubing, S.A.E. 1015, is perhaps most representative. A sample was reduced from 1⅛ inch O.D., #10 (.134 inch) wall to 1-inch O. D., #13 (.095 inch) wall, (35½ per cent reduction) in one pass, using a short or stationary mandrel. The tube was first annealed by the usual process, namely heated to 1400 degrees Fahr., saturated, and freely cooled, all in controlled neutral atmosphere. It was lightly pickled and lubricated in "dope." The tube was drawn part way through the die, stopped and backed out, giving a part near the center illustrating the process of reduction on entering and passing through the die. This portion was sectioned longitudinally, one side deep etched for grain flow and the other examined for microstructure. (The cut removed about $\frac{1}{32}$ inch of metal on each side of centerline.)

Hardness readings were made and tensile specimens taken on each end, representing conditions before and after drawing. Dimensions and results of physical tests are given in Table I. It is of special interest to compare the yield loads and breaking loads of the drawn and undrawn sections. These are each about 400 pounds greater after drawing than before. The yield stress and ultimate stress have been raised by cold work just a little more than enough to offset the reduction in area produced by cold drawing.

Presumably, the maximum pull in pounds which may be applied by the draw head is less than the yield load (in pounds) of the

Table I
Seamless Steel Tube (S.A.E. 1015) Before and After Cold Drawing

1⅛ Inch to 1 Inch O.D. Single Pass. Short Mandrel. Reduction 35.5 Per Cent

	Before	After
Outside Diameter, Inches	1.125	.998
Wall, Inches	.133	.094
Area, Square Inches	.414	.267
Yield Load, Pounds	14,200	14,600
Ultimate Load, Pounds	23,700	24,100
Yield Stress, P.S.I.	34,400	54,600
Ultimate Stress, P.S.I.	57,400	90,500
Elongation in 2 Inches, Per Cent	55	16.5
Rockwell Hardness	B70	B95

drawn tube. If the force required to cause reduction is equal to the yield load of the undrawn tube, then the force available and necessary to overcome die and mandrel friction is, in the present instance, less than 400 pounds. This seems a small figure, and if true, illustrates the extreme importance of adequate lubrication. If, on the

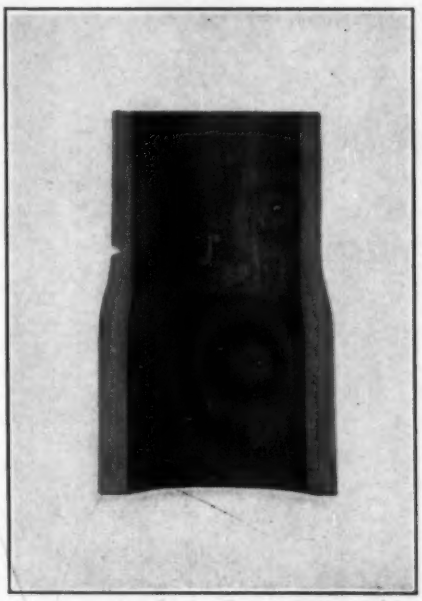

Fig. 6—Photograph of Longitudinal Section, Sample Tube, Moderately Hot-Etched, Nearly Full Size.

other hand, the frictional pull is greater than indicated, then the force required to cause reduction must be less than the yield strength of the undrawn tube.

Flow lines were not revealed by ordinary hot etching. The sample was evidently rather free from flaws or segregations. See Fig. 6.

Examination of the profile at the reduced section shows that a large amount of reduction in diameter has taken place by pulling inward before coming in contact with the die or mandrel. Measurable reduction began about half an inch ahead of the point of first die contact.

Microstructure

The effect of (35 per cent) reduction on the microstructure was pronounced, but not extreme. See Fig. 7. There was no marked

difference in structure from the outer surface to the inner surface in any plane, except for the usual slight amount of surface decarburization.

The photomicrographs were taken half way between the inner

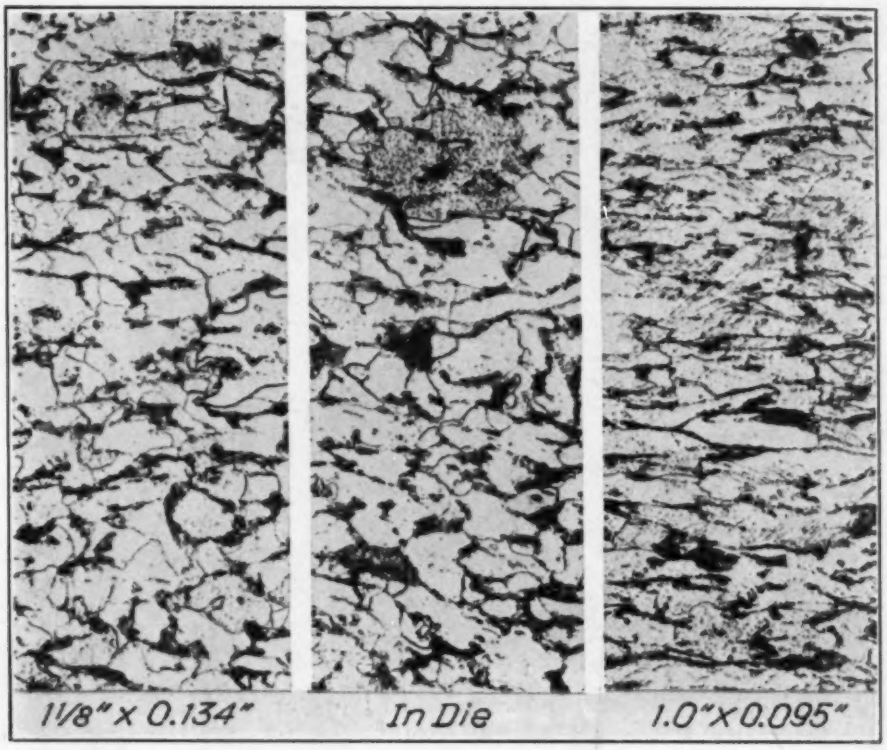

Fig. 7.—Photomicrograph of Sample Tube. Left Photomicrograph is Longitudinal Section. Center Photomicrograph, Before Entering Die. Right Photomicrograph, After Leaving Die. Axis Horizontal.

and outer tube surfaces at the beginning, middle, and end of the reduced section of the wall, spanning total length of about 8 millimeters or 0.32 inch.

Production Factors

The cold drawn tube industry exists by virtue of the limitations of hot working as to diameter, wall thickness, finish, and shape. Strength, produced by cold work, is also a factor.

The purpose of the cold drawing mill is therefore to reduce the relatively large and crude hot-rolled tube to the desired final dimensions as economically as possible. The amount of metal worked per unit of time is determined fundamentally by:

A. Reduction of area per pass, in per cent.
B. Speed of pull, feet per minute.

These factors depend upon:

1. Physical properties of the material,
   (a) Before drawing, as affected by
      (1) Composition
      (2) Annealing treatment (grain size, carbides).
   (b) During drawing, as affected by
      (1) Cold work (plastic flow)
      (2) Effect of heat caused by drawing.
2. Die and mandrel forms for efficient flow of material.
3. Lubrication (including metallic and nonmetallic coatings, etc.).

There is also a group of incidental factors which have a great influence on output, principally as causes of delay:

4. Incidental
   (a) Variations in composition of material, within specified tolerance.
   (b) Mechanical defects and irregularities.
   (c) Variations in annealing treatment affecting
      (1) Physical properties
      (2) Surface condition
      (3) Decarburization, etc.
   (d) Wear of dies and mandrels as affected by their
      (1) Composition
      (2) Heat treatment
   (e) Lubrication troubles, including
      (1) Surface condition of tubing, presence of scale, dirt, etc.
   (f) Time lost in changing tubes.
   (g) Time lost in renewing dies and mandrels.

Factors 1, 2, and 3 have to do with the basic laws of cold drawing, while those under group 4 relate to quality control, engineering, and management. All deserve careful attention by the technician, aided (not hindered) by the shop man's experience.

Limitations of Cold Drawing

Experiments conducted some years ago by the writer with a process of lead coating, in which the lead was truly alloyed with the

surface of the tube and therefore adhered, showed that mild carbon steel could be given as many as ten successive passes at about 35 per cent reduction per pass, a total reduction of about 97 per cent of the original area, without re-annealing, whereas two passes were the maximum obtainable with ordinary lubrication and the same reduction per pass. The steel ceased to increase in hardness after the first few passes. It could be drawn successfully even in this severely cold-worked condition, but would have ruptured due to frictional pull without the superior lubrication afforded by the lead.

Unfortunately, the data obtained in these experiments are not now available. But it is clear that the limitations of cold drawing of thin walled tubing reside in surface friction, rather than in the metal itself and that the possibilities for economy in this direction are by no means exhausted.

THE COLD ROLLING OF MILD STEEL SHEETS AND STRIP

By Anson Hayes and Robert S. Burns

Abstract

The authors have dealt only with the more fundamental aspects of the problem in connection with the metallurgical phases of the discussion. Theoretical discussion includes a simplified treatment of the mechanism for reduction by rolling, with particular reference to the so-called no-slip point. Data are given to show the effects of both front and back tension on the position of the no-slip point, and the effects of such tension on the metallurgy of the product and the power required to make reductions. Stress-strain curves show the effects of various amounts of cold reduction on the shape of such curves both before and after annealing. Correlations among the amount of yield point elongation, yield point in pounds per square inch, and the grain size are shown by these curves and the photomicrographs. Under the subject of Finishing, the elimination of yield point elongation is shown to be a function of the initial amount of yield point elongation, the diameter of the rolls used for tempering, and the surface of both the sheet and the rolls. A discussion of both roller leveling and stretcher leveling is included.

THE authors have undertaken to discuss those parts of the broad outline for the Symposium on the Plastic Working of Metals devoted to types and applications of equipment, reasons for their use, flow of metal, fibre and grain size resulting from cold rolling operations, and these same features for cold straightening processes. Because of the very broad field that would be involved and the length of a paper dealing with such a wide scope, we have chosen to discuss these subjects as are involved with mild steel sheets and strip. To further limit this discussion, we have chosen to deal with a few types of equipment being used most at the present time. We have dealt with only the more fundamental aspects of the metallurgical phases of this one product—mild steel.

A paper presented as part of the Symposium on the Plastic Working of Metals, Eighteenth Annual Convention of the Society held in Cleveland, October 19 to 23, 1936. Of the authors, Anson Hayes is Director, Research Laboratories, and R. S. Burns is Research Engineer, Research Laboratories, The American Rolling Mill Co., Middletown, Ohio. Manuscript received June 24, 1936.

As rolling operations are carried out in practice, cold rolling sheets and strip material at essentially room temperature is a simpler process with which to deal metallurgically than is true of the hot rolling process. As most commonly practiced, hot rolling is carried out at temperatures where two opposing processes are in action: One, the deformation of the metal which develops what is usually designated as "strain," and that of annnealing, which has the effect of reducing the strain which deformation by the rolls is producing. The resulting strain in a hot-rolled material may, therefore, be considered as the net result of the development of strain through deformation and that of the annealing effect, which may be a simple relief of some of the strain, or of an actual partial recrystallization. Complete recrystallization, as is well known, removes much more completely the effects of severe deforming processes than do annealing operations at temperatures which are too low and for lengths of time which are too short to bring about such changes in structure.

There has been from time to time much discussion regarding the definition of hot and cold rolling of sheets and strip. Badlam[1] has defined cold rolling as "the working or shaping of a material by means of rolls at temperatures below the critical range." He defines hot rolling as "working above the critical range."

In the light of what has just been said, the authors believe that a better definition can be formulated. The cold rolling operation carried out as it is near room temperatures, leaves essentially the fully strained structure produced by the deformation of rolling. The process of rolling at temperatures in the neighborhood of 1100 to 1250 degrees Fahr., while well below the critical range, has the strain condition produced by rolling very markedly modified by the annealing effect. Viewed from the standpoint of the structural effects of the rolling process, it would seem that a better definition of cold rolling would be "rolling at temperatures sufficiently low as to result in essentially no change due to annealing of the deformed structures which result from the reduction of gage by rolling."

Historical

Records show that cold rolling mills were used as early as 1550 for the production of bars of gold and silver. According to Badlam,[2]

[1]Yearbook of the American Iron and Steel Institute, page 352 (1930).
[2]Loc. cit.

the beginning of the cold rolling of iron and steel as an industry may be said to date from about 1830, as at that time we find records of the manufacture and use of steel reeds by the weavers of Barmen and Hohenlimburg in Germany. These reeds were made of high carbon steel, flattened by cold rolling in exactly the same manner as is practiced in the flat wire industry today. They, as well as watch and clock springs, probably constituted the production of that period. Krupp had started the manufacture of hardened steel rolls for cold rolling steel at about this same time.

When one considers the large tonnages of cold-reduced strip and sheet steel consumed at the present time, the historical sketches of Badlam are interesting in that they show the chief uses of cold-rolled steel have risen from products such as wire for ladies' hats, hoops for the old-style hoop skirts, corset staves, clock springs, and shank steel for shoes, to sheets and strip of width of 90 inches or more with excellent surface and physical properties made in tonnages never dreamed of in that earlier period. For a most interesting detailed chronological story of the gradual process of development of the necessary elements in equipment and its operation for manufacturing cold-rolled steel and the development of markets for it, Badlam's article,[3] together with that of J. R. Adams,[4] is very good. In these references is given a clear conception of the struggle in the development of materials, machine designs, and methods of applying power, which are characteristic of our present-day steel mill units. The earliest mills were driven by water power. These were supplanted by steam. The steam engine, in turn, gave way to the use of constant-speed motors, which today have been supplanted by the highly flexible individual direct current motor.

Developments in materials used for mills and for annealing and handling the cold-rolled material are no less phenomenal. Today it can be said that our present cold rolling methods of manufacturing mild steel sheets and strip are the result of a large number of contributing factors in scientific progress, in improvement of materials, and in experience.

Theoretical

At the present time, equipment for cold reducing stock in the manufacture of steel sheets and strip is available, which makes use

[3] Loc. cit.

[4] Yearbook of American Iron and Steel Institute, page 115 (1924).

of one, or combinations of two or more, of the following methods of applying power to bring about the thinning of section:

1. All power applied to the working rolls.
2. All power applied as a combination of forward and back tension.
3. Part of power applied to rolls and part as forward tension.
4. Part of power applied to rolls, part as forward tension, and part as back tension.

In the following simplified treatment of the mechanism for reduction by cold rolling, attention will be confined to actions going on between the point of entry of the stock into the rolls and that of exit. So far as the roll surfaces are concerned, this region will be called the "arc of contact."

Under the most usual conditions for rolling, there is a point within the region mentioned above for which the horizontal component of the forward movement of the stock is equal to that of the roll surface. This point will be called the *no-slip point*. Back of this point, the stock has a forward horizontal component of velocity less than that of the roll surface. In front of it, the stock has a greater forward horizontal velocity component than the roll surface. This means that back of the no-slip point, there is backward slip of the stock relative to the roll surface; while in front, there is forward slip. The rolling process of reducing the thickness of metal may, in fact, be considered as a special extrusion process.

The location of the no-slip point depends upon a number of factors, such as:

1. Percentage of reduction effected.
2. Physical properties of the metal under reduction.
3. Combination of methods of applying power, as indicated earlier, and its distribution between the various methods.
4. Coefficient of friction between rolls and stock.

In the analysis which follows, the principal simplifying assumptions made are:

1. No side spread in the stock during reduction.
2. No change in density of material.
3. The horizontal component of forward velocity is constant across the section of the stock at any point in the arc of contact.

4. The coefficient of friction between the stock and the roll surface is constant at any one point within the arc of contact.
5. Elastic deformation is negligible as compared to plastic.

In Fig. 1

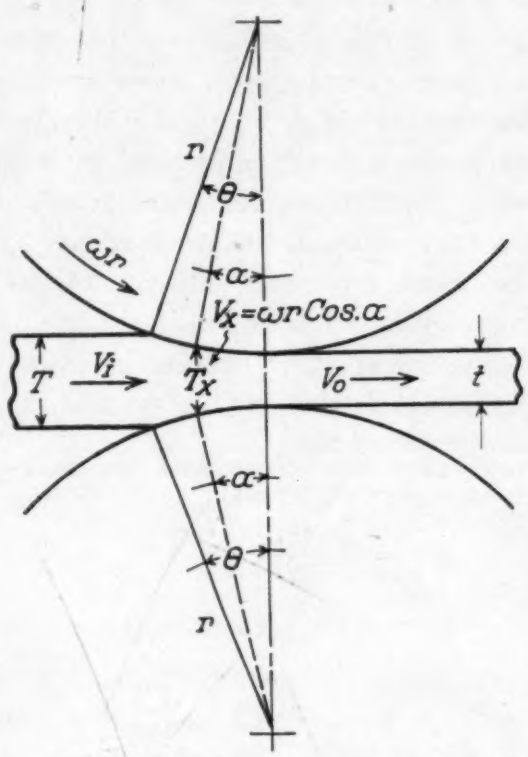

Fig. 1—Diagram of the Process of Reduction by Rolling.

Let θ = angle of contact between rolls and stock.
α = angle to the no-slip point.
r = radius of rolls.
ω = angular velocity of rolls.
T = thickness of incoming stock.
t = thickness of outgoing stock.
t_x = thickness of stock at no-slip point.
V_i = horizontal component of velocity of incoming stock.
V_o = horizontal component of outgoing stock.
V_x = horizontal component of stock at the no-slip point.

Equal volumes of stock are passing the various points in the arc of contact in the same interval of time—consequently:

$$(1)\quad V_i \frac{T}{t} = V_o$$

$$(2)\quad V_x = \frac{t}{t_x} V_o$$

(3) $V_x = \omega$ r cos α since this is the no-slip point.

(4) $t_x = t + 2r\ (1 - \cos \alpha)$

(5) $T = t + 2r\ (1 - \cos \theta)$

By combining (2), (3), and (4)

(6) $$V_x = \omega r \cos \alpha = \frac{tV_o}{t + 2r\ (1 - \cos \alpha)}$$

From (6) α can be computed from r, ω, V_o, and t.

However the use of this equation is somewhat inconvenient for computational purposes. Fortunately, there are simple methods of experimental procedure by which V_x can be obtained.

Assume a scribe mark is made parallel to the axis of the roll. The circumference is the distance between the imprints of this mark on the stock, if no elongation on the stock occurs after it passes the no-slip point. The distance between any two marks continues to increase up to the completion of reduction in the arc of contact. However, the same volume of stock is present between these two scribe marks at all points from the no-slip point to exit.

If d_x = circumference of the roll,
d = distance after exit between any two marks; then
$t_x d_x = td$ = volume of stock; and

(7) $$\frac{t}{t_x} = \frac{d_x}{d}$$

From (2) and (7)

(8) $$V_x = \frac{d_x}{d} V_o$$

And from (6) and (8)

(9) $$V_x = \frac{d_x}{d} V_o = \omega r \cos \alpha = \frac{tV_o}{t + 2r\ (1 - \cos \alpha)}$$

Equation (9) may be used to compute cos α from the issuing velocity of stock, the linear velocity of the roll surface, and the ratio of the roll circumference and the distance between any two adjacent marks on the stock after exit. Further consideration shows that forward slip, S_o, between roll and stock at the point of exit is given by

(10) $S_o = V_o - \omega r \cos \alpha$

Horizontal component of back slip at entry, S_i, is given by

(11) $S_i = \omega r \cos \alpha - V_i$.

To get circumferential slip between roll surface and stock at entry, consideration of geometry of Fig. 1 gives

(12) $\frac{V_i}{\cos \theta}$ = Tangential velocity of stock at entry, and, consequently

(13) $\omega r - \frac{V_i}{\cos \theta}$ = circumferential back slip between surface of stock and roll face at entry.

To show the effects of both front and back tension on the position of the no-slip point, the following data were obtained in experi-

ments carried on with a mill where both front and back tension could be measured accurately and varied over a wide range. To calculate alpha, the following equation (9) was used:

$$\cos \alpha = \frac{tV_x + 2rV_x - tV_o}{2rV_x}$$

Table I

Effect of Variations in Front and Back Tension on the Exit Velocity and Angle Alpha with Constant Roll Speed

Experiment #	Gage In	Gage Out	% Reduction	Front #/sq. in.	Back #/sq. in.	V_o	V_x	Alpha
	Roll Radius = 2.077					cos Theta = .99037		Theta = 7°58′
1	0.067	0.047	29.8	0	0	13.513	13.050	2°18′
2	0.067	0.0465	30.6	26500	0	14.270	13.050	3°44′
3	0.067	0.047	29.8	26500	18600	13.694	13.050	2°43′
4	0.067	0.047	29.8	0	0	13.513	13.050	2°18′
5	0.067	0.047	29.8	10000	0	13.658	13.050	2°38′
6	0.067	0.047	29.8	22000	0	14.060	13.050	3°24′
7	0.067	0.047	29.8	30000	0	14.435	13.050	3°58′
8	0.067	0.047	29.8	42500	0	15.200	13.050	4°57′
	Roll Radius = 2.077					cos Theta = .99182		Theta = 7°20′
9	0.034	0.017	50.0	40000	28600	14.655	13.050	2°34′
10	0.034	0.017	50.0	40000	25700	14.694	13.050	2°36′
11	0.034	0.017	50.0	40000	22800	14.700	13.050	2°36′
12	0.034	0.017	50.0	40000	20000	14.759	13.050	2°39′
13	0.034	0.017	50.0	40000	17100	14.890	13.050	2°45′
14	0.034	0.017	50.0	40000	14300	14.970	13.050	2°49′
15	0.034	0.017	50.0	40000	11400	15.112	13.050	2°55′
16	0.034	0.017	50.0	40000	8550	15.269	13.050	3° 1′
17	0.034	0.017	50.0	40000	0	15.699	13.050	3°18′

Experiments 1, 2 and 3 (Table I) were intended to show the effect of no tension, front tension only, and equal pull front and back (not equal tensions front and back because strip is 30 per cent heavier on entrance side of mill; hence, unit stress is 30 per cent less.) An attempt at rolling with no front tension and 18,600 pounds per square inch back tension was futile, because the rolls would not pull the strip through under this high back tension, and roughening of the rolls would have altered the conditions of the test.

Experiments 4 through 8 show the effect of variations in front tension when no back tension is applied. Note that the speed of the outcoming strip varies from 13.5 to 15.2 with the same roll speed. This means that the equilibrium between forward and backward slip has been radically changed by the application of this high front tension. Experiments 9 through 17 were made under the same mill conditions so far as roll speed and roll surface are concerned, but the use of lighter gage stock and a heavier reduction (50 per cent) are involved. These data show that the position of the no-slip point, and, hence, the equilibrium between front and back tension, may also be altered by changes in back tension.

Insofar as the metallurgical and X-ray structures of the deformed metal is concerned, the X-ray diffraction patterns and micrographs shown in Figs. 2 and 3, respectively, were made on mild steel cold reduced 83 per cent, with and without tension, in three passes as follows:

Pass	Gage In	Gage Out	Per Cent Reduction	Tension Out Lb./Sq. In.
1	0.067	0.035	47.1	25,500
2	0.035	0.017	51.4	28,400
3	0.017	0.011	35.3	35,000

The samples cold reduced without tension were made in the same steps of reduction. The X-ray patterns show no differences in the degree of preferred orientation produced by rolling with tension, and the photomicrographs show no differences in structure caused by tension rolling, either before or after annealing. Many other annealing treatments were performed on these samples, but no differences were found in rolling with and without tension.

No definite and concise information has come to our attention that shows any difference metallurgically caused by rolling all in one direction or by reversing the rolling direction each pass.

Experiments have shown that as the front tension applied is increased, the net energy necessary to reduce the strip is decreased. These tests, however, were made with a constant differential pull of 300 pounds forward per inch of strip width. Hence, as the front tension was increased from 700 to 1400 pounds per inch of strip width, the back tension was increased from 400 to 1100 pounds per inch of strip width. This increase in front and back tension decreases the screw pressure about 30 per cent, which in turn decreases the power lost as friction between the roll face and the strip. Therefore, if we consider only the net energy required to deform the metal, and disregard the higher loss of power at the brake, or economically retrieve this power, then the energy required to deform the strip decreases with higher front and back tensions. However, if the power lost at the brake is considered as "lost," the power required to deform the strip increases as the front and back tension are increased.

With increasing back tension (front tension constant at 40,000 pounds per square inch), the total energy required to reduce the strip increases because of the higher power losses at the brake. If the power lost at the brake could be recovered economically, then operation with a high back tension would be economical. Screw pressure decreases with increasing back tension, but changes in front

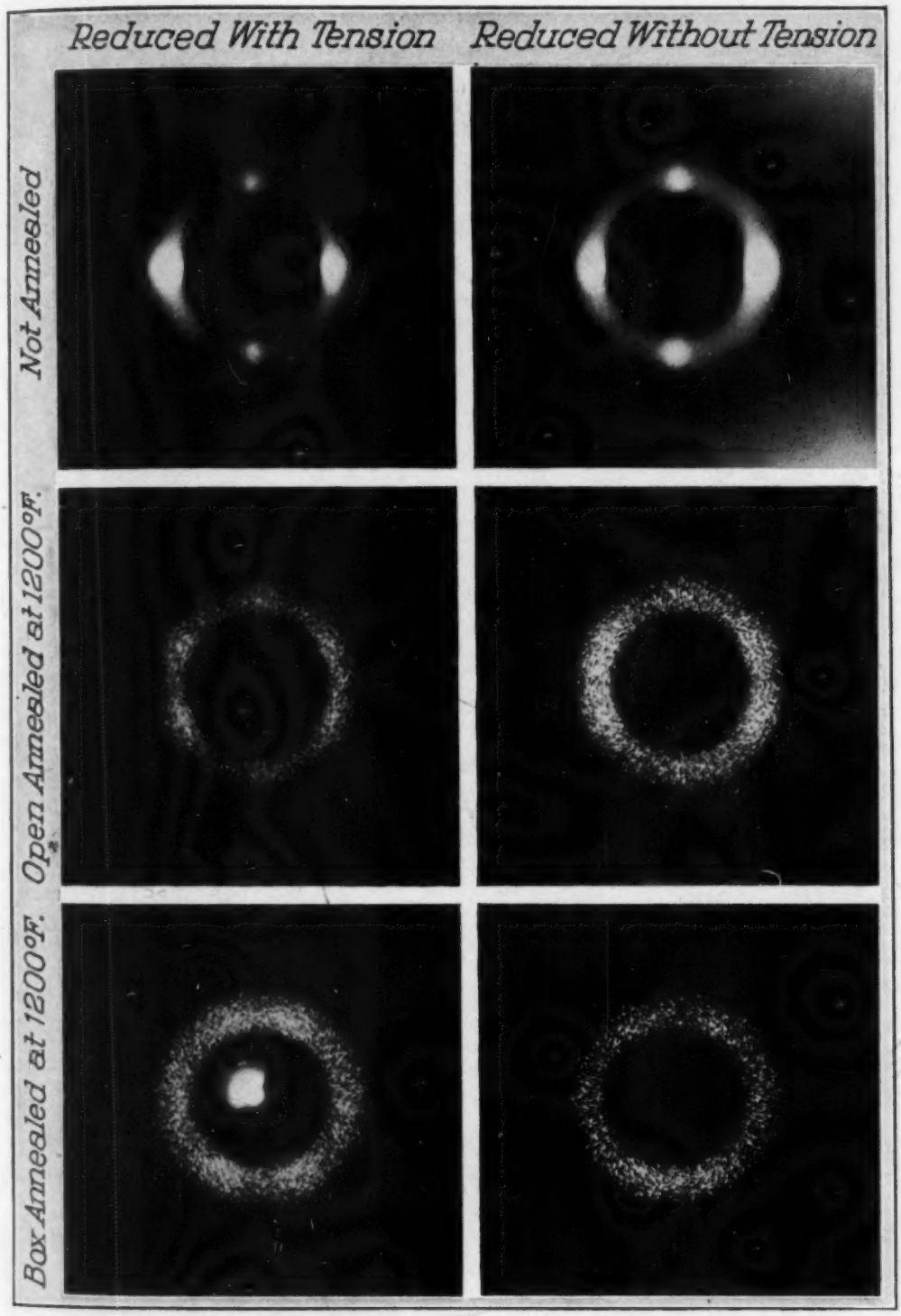

Fig. 2—X-Ray Diffraction Patterns of Mild Steel as Cold Reduced and After Cold Reducing and Annealing at 1200 Degrees Fahr. Samples Cold Reduced 83 Per Cent With and Without Tension. a—Reduced With Tension, Not Annealed. b—Reduced Without Tension, Not Annealed. c—Reduced With Tension, Open Annealed at 1200 Degrees Fahr. d—Reduced Without Tension, Open Annealed at 1200 Degrees Fahr. e—Reduced With Tension, Box Annealed at 1200 Degrees Fahr. f—Reduced Without Tension, Box Annealed at 1200 Degrees Fahr.

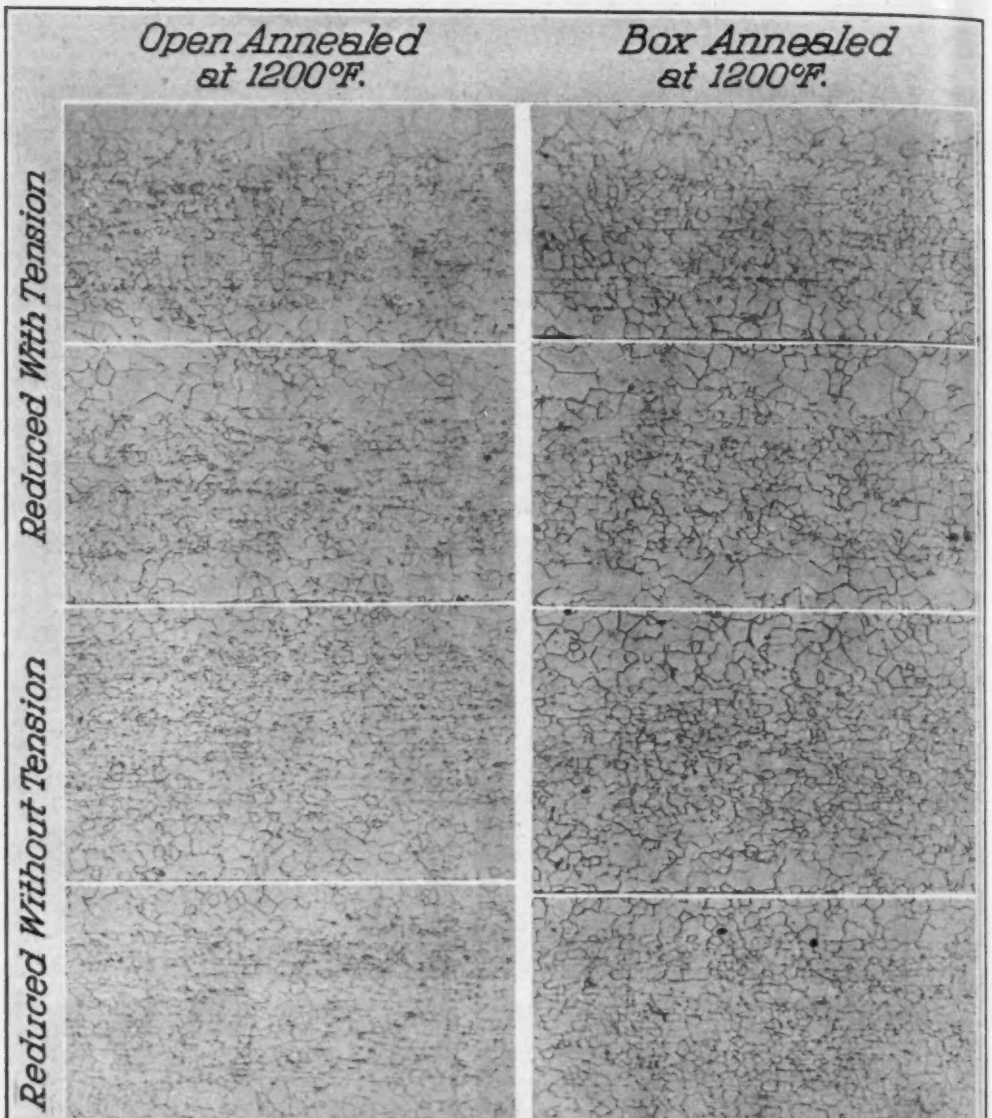

Fig. 3—Micrographs of Mild Steel. Cold Reduced 83 Per Cent With and Without Tension and Annealed at 1200 Degrees Fahr. a—Reduced With Tension, Open Annealed at 1200 Degrees Fahr. b—Reduced Without Tension, Open Annealed at 1200 Degrees Fahr. c—Reduced With Tension, Box Annealed at 1200 Degrees Fahr. d—Reduced Without Tension, Box Annealed at 1200 Degrees Fahr. Etched in 3% Nital. × 100.

tension with constant back tension seem to have little, if any, effect on the screw pressure.

It is obvious from the foregoing that there may be combinations of the percentage of power supplied to the mill or reel which give most economical results. These conditions depend so largely upon the physical characteristics of the product and the mill at hand that each operator has varied the conditions to suit his particular problem.

Insofar as smaller rolls decrease the screw pressure, it would be expected that decreasing the roll size would decrease the power lost as friction between the strip and the roll face, and hence, decrease the net energy required to accomplish a given reduction. Small rolls, however, must be reground oftener than large rolls.

In the theoretical treatment of an earlier section on the size of the no-slip angle in cold rolling, the assumption was made that the horizontal component of forward progress of all parts of the cross

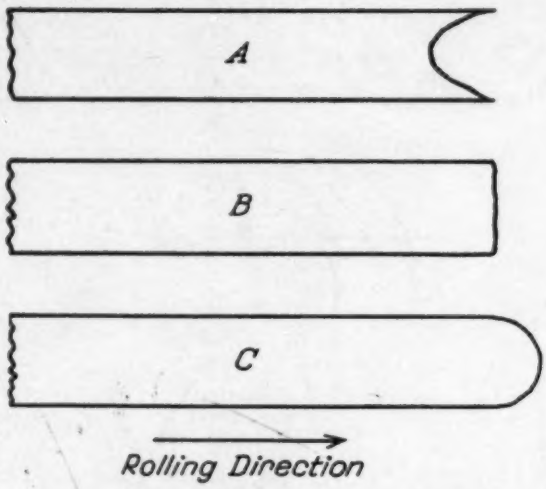

Fig. 4—The Shape of the Front End of the Stock as Varied by Light and Heavy Reductions.

section of the stock at any one point within the arc of contact is the same. Experience has shown that this assumption is true in only very special cases. In Fig. 4 are shown three types of fronts which are obtained.

For very light cold reductions, such as are used in temper rolling, the front of the stock takes the form shown in Fig. 4A. The concave form of this front shows that the surfaces of the stock slip forward faster than do points in the stock farther from the surface. In heavy reductions, a convex front is obtained, as shown in Fig. 4C. For some intermediate reductions between those results in the shape shown in Fig. 4A and that in Fig. 4C, a straight front may be obtained as shown in Fig. 4B.

Cold Rolling Equipment

The type of cold mill used for a particular product depends to a large extent on the desired characteristics of the product and the equipment available at the time of the installation. Stephen Badlam

has discussed these points very ably in his papers in the Yearbook of American Iron and Steel Institute, 1930, and the December, 1935, issue of Iron and Steel Engineer. The four types of cold reduction mills which will be mentioned are: (1) tandem mill, (2) single stand reversing mill, (3) Steckel mill, (4) cold sheet mill. The advantages of the tandem mills for reducing coils are: (1) great capacity, (2)

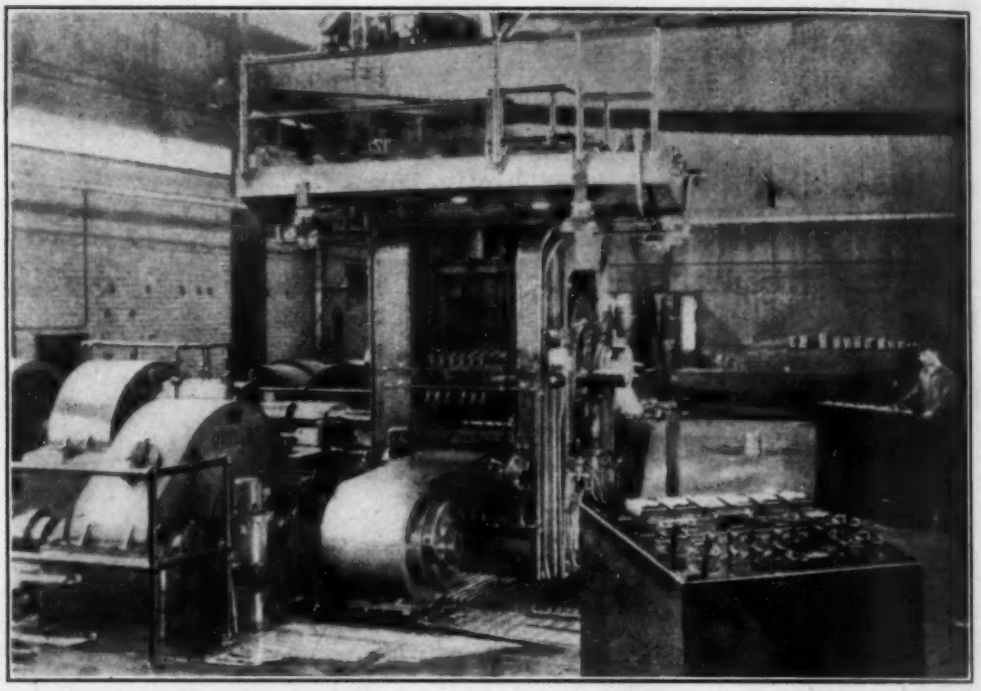

Fig. 5—Single Stand Reversing Mill.

low labor cost, although the installation cost of the tandem mill is very high. The disadvantages of the tandem mills involve lack of flexibility for rapid section changes and the fact that the stock must receive a fixed and limited number of passes.

With the single stand reversing mill (Fig. 5) less floor space is required than for the tandem mill. The single stand reversing mill also has great flexibility as to the number of passes, a lower initial cost than the tandem mill, and a slight possibility of better gage control. The labor cost for the single stand reversing mill is somewhat high, because such a mill is limited in capacity.

The Steckel mill shown in Fig. 6, although relatively inexpensive, has a high initial cost per ton of output. The Steckel mill is like the single stand reversing mill, in that it is flexible as to number of passes, and that it occupies a small amount of floor space. The

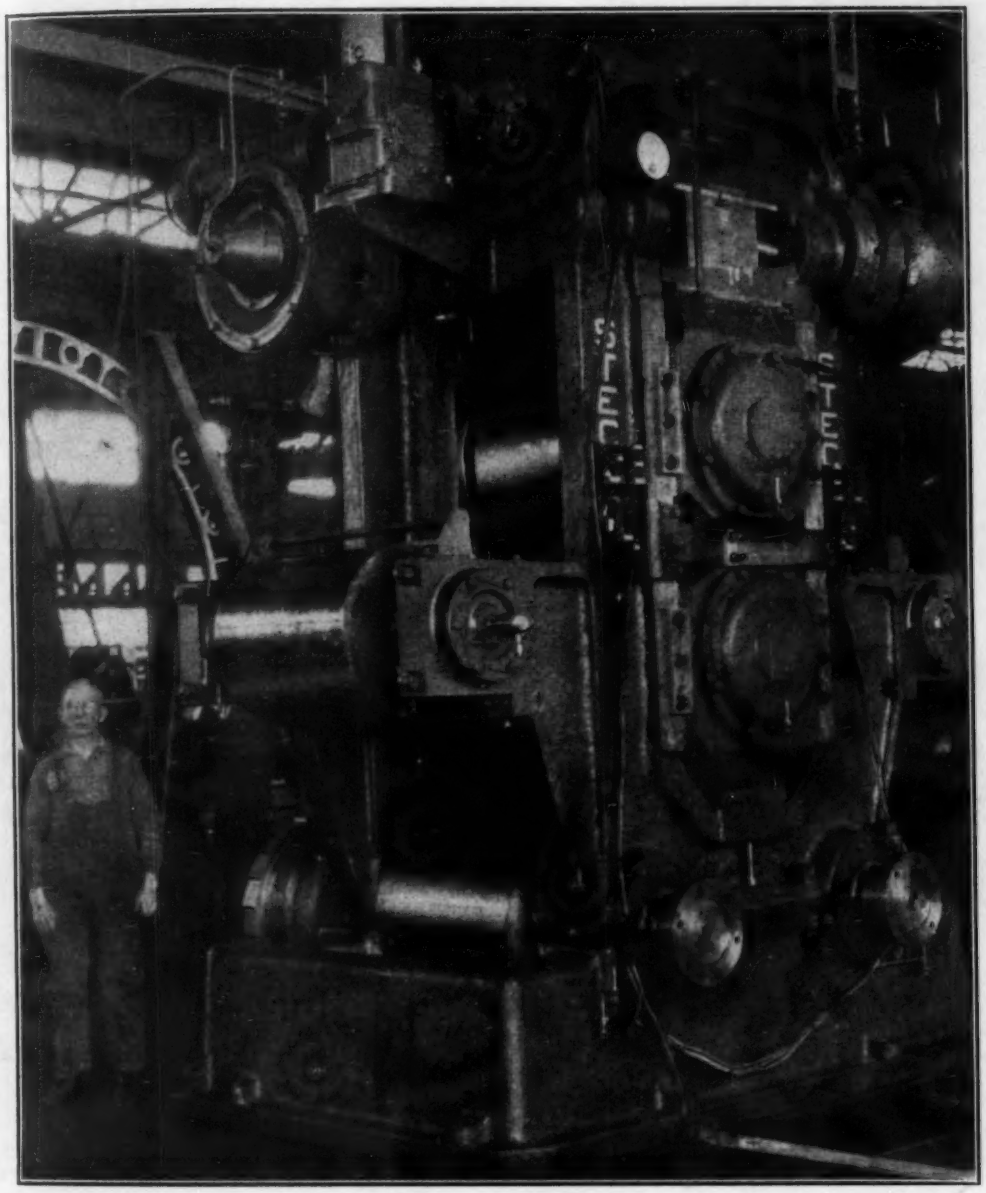

Fig. 6—Steckel Mill.

Steckel mill has been classified as a device for rolling only the lightest gages of mild steel. However, this mill is also a useful tool in the rolling of tough stainless steels, high carbon steels, and razor-blade stock, regardless of finishing gage. The Steckel mill may be classed as a production unit, if by the use of such a mill two or more intermediate anneals may be eliminated. The capacity of the Steckel mill is approximately one-third that of the tandem mill and about one-half that of the single stand reversing mill. There is also a high scrap loss because of the end scrap on both ends of each coil.

The cold sheet mill, such as the installation at the Ashland Works of the American Rolling Mill Co., a photograph of which is shown in Fig. 7, was installed to utilize breakdowns where coiled strip was not available. The output of such a mill may exceed 50

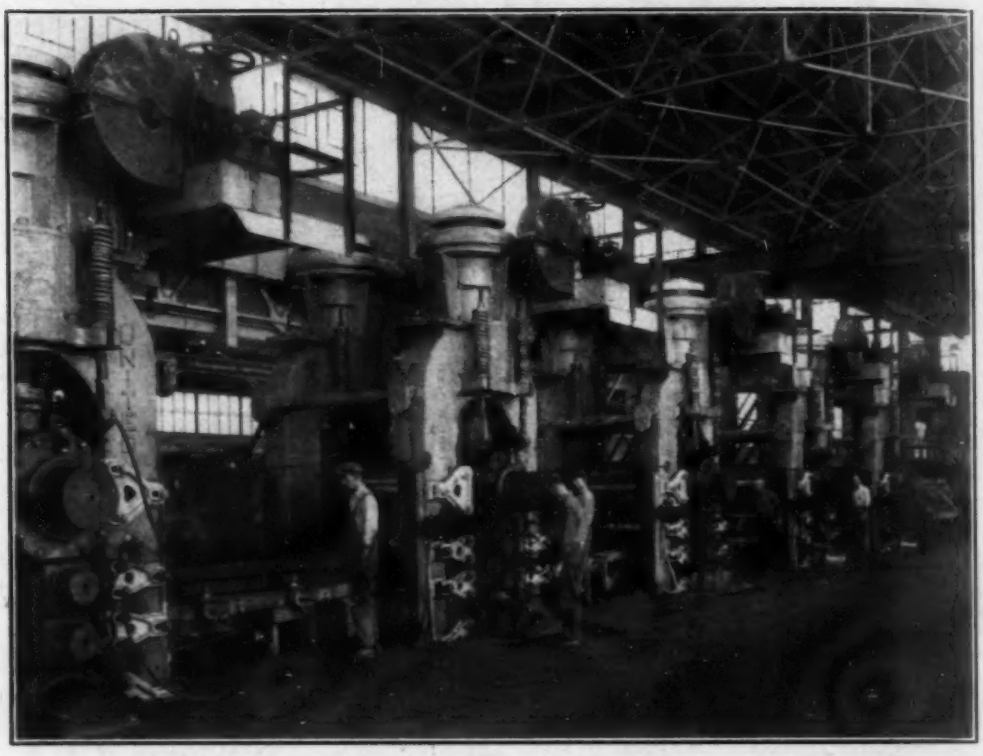

Fig. 7—Tandem Mill for Cold Reducing Sheets at the Ashland Works of The American Rolling Mill Company.

tons per hour in the wider widths. The difficulties encountered with the use of such a mill are lack of flatness in the product and low economy on account of end scrap on each sheet.

There are many other advantages and disadvantages associated with each of these mills. There are also other types of special mills, such as the ring mill and Sendzimir mill, which may or may not have practical applications.

Effects of Cold Reduction on Cold-Rolled Structures and Properties and on Annealed Structures and Properties After Box Annealing

In Fig. 8a is shown the various degrees of distortion of the microstructure accompanying cold reductions from 0 to 60 per cent, and in Fig. 9, taken from a paper by R. L. Kenyon and R. S.

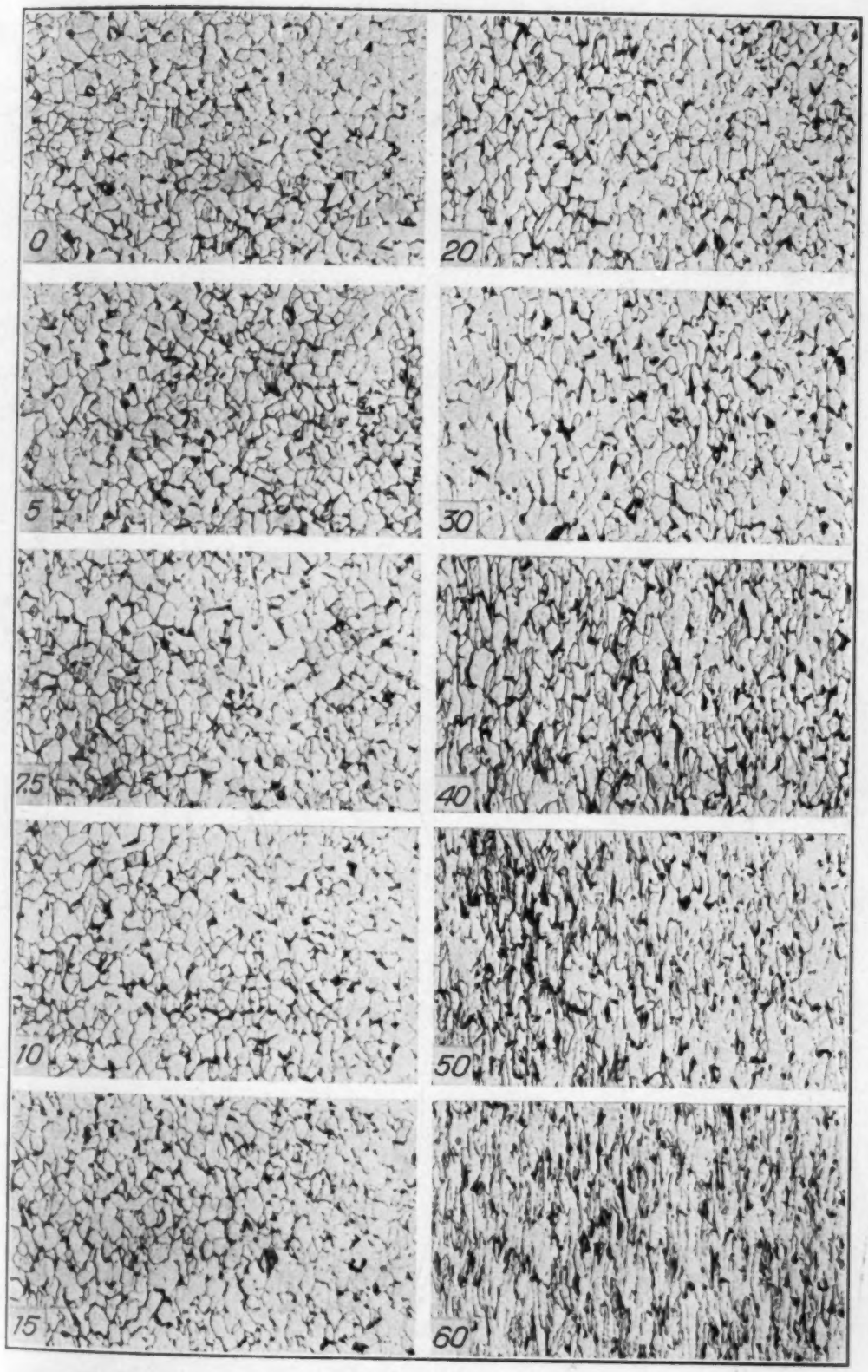

Fig. 8a—Microstructure of Mild Steel Normalized and Cold Reduced Various Amounts. Etched in 3 Per Cent Nital. × 100.

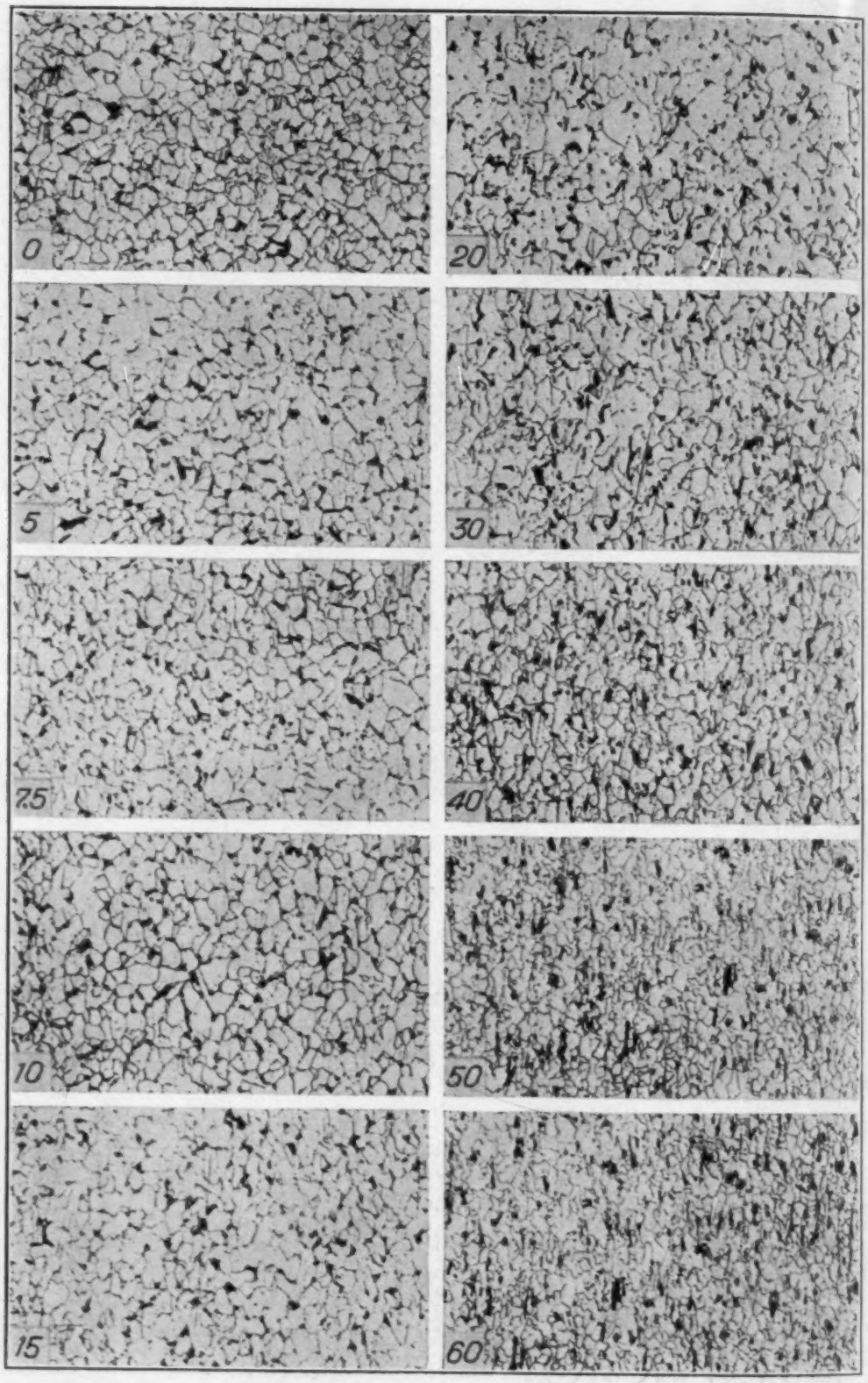

Fig. 8b—Microstructure of Mild Steel Cold Reduced Various Amounts and Box Annealed 4 Hours at 1100 Degrees Fahr. and Cooled 25 Degrees Fahr. per Hour. Etched in 3 Per Cent Nital. × 100.

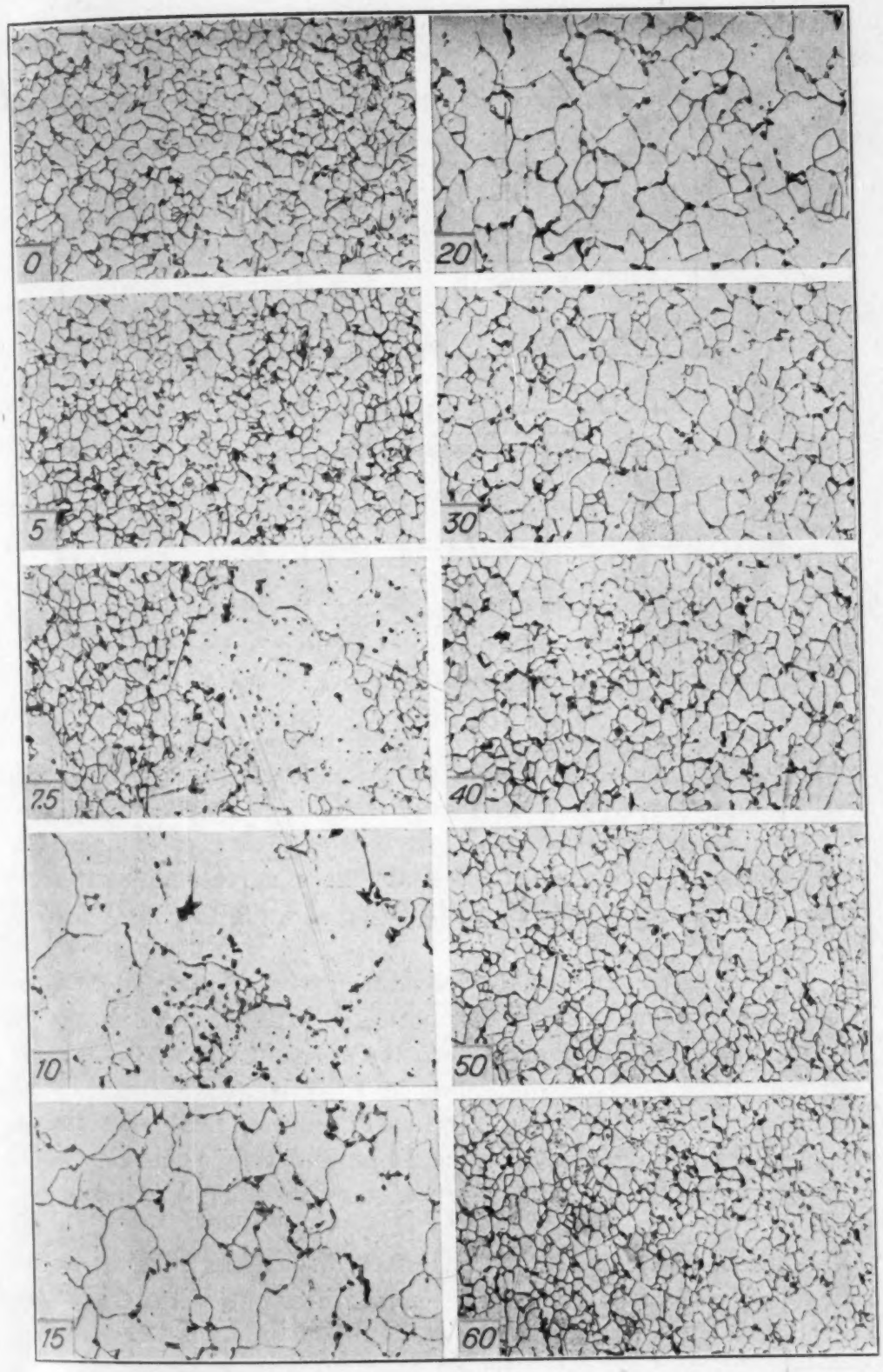

Fig. 8c—Microstructure of Mild Steel Cold Reduced Various Amounts and Box Annealed 4 Hours at 1350 Degrees Fahr. and Cooled 25 Degrees Fahr. per Hour. Etched in 3 Per Cent Nital. × 100.

Burns,[5] are shown the effects of various percentages of cold reduction on the shape of the stress-strain curves for cold-rolled material. In Fig. 10 is shown the relation between Rockwell hardness and the

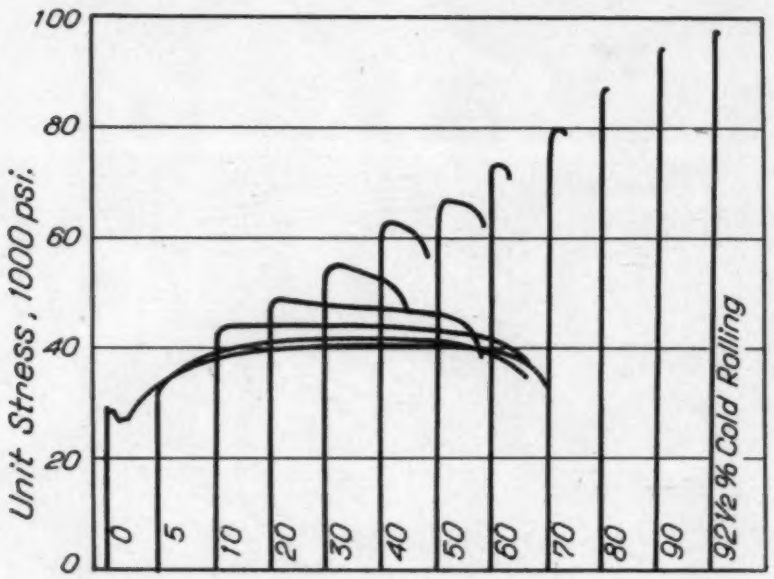

Fig. 9—Effect of Cold Reduction on the Shape of the Stress-Strain Curve of Mild Steel Sheets.

amount of cold reduction for the same samples shown in Fig. 8. In Fig. 8b are shown the various grain structures obtained after box annealing these same cold-rolled materials at 1100 degrees Fahr. In Fig. 11 (above) are shown stress-strain curves corresponding to various amounts of cold reduction after box annealing at 1100 degrees Fahr.

In Fig. 8c are shown the microstructures of material cold reduced various amounts and box annealed at 1350 degrees Fahr.

In Fig. 11 (below) are shown stress-strain curves for various cold reductions after box annealing at 1350 degrees Fahr. Fig. 12 includes graphs of variation in Rockwell hardness, yield point, tensile strength, per cent elongation in 2 inches, and yield point elongation after box annealing as affected by various percentages of cold reduction.

Some of the more important effects of various percentages of cold rolling and of box annealing temperatures on the material after box annealing are as follows:

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[5] "Autographic Stress-Strain Curves," TRANSACTIONS, American Society for Steel Treating (1932).

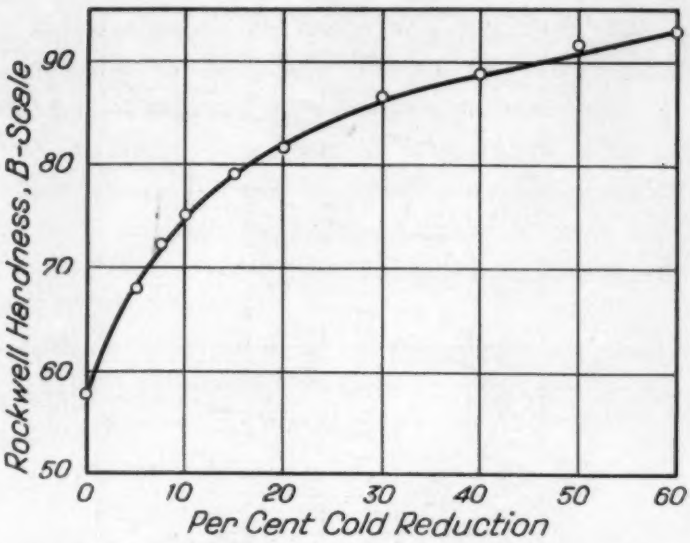

Fig. 10—Effect of Cold Reduction on the Rockwell "B" Hardness of Mild Steel Sheets.

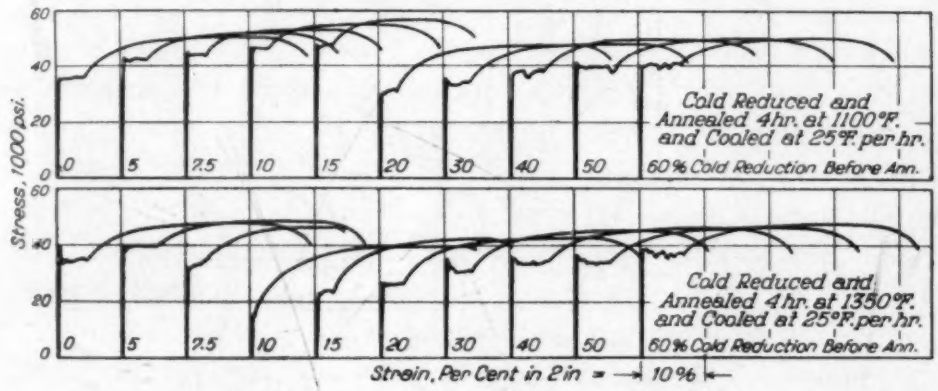

Fig. 11—Effect of the Amount of Cold Reduction Before Annealing on the Shape of the Stress-Strain Curve for Mild Steel After Box Annealing at 1100 and 1350 Degrees Fahr.

1. There is a degree of strain which is critical in its effect on grain size and physical properties for each box annealing temperature. The results of this critical strain, with that of the box annealing temperature, are to produce large grain size with low yield point, low Rockwell hardness, low tensile strength, and lower elongation.
2. Beyond the critical strain region, the grain size is reduced as the percentage of cold reduction increases for a given box annealing temperature. This decrease in grain size is accompanied by increase in yield point, increase in yield point elongation, increase in tensile strength, and increase in total elongation.

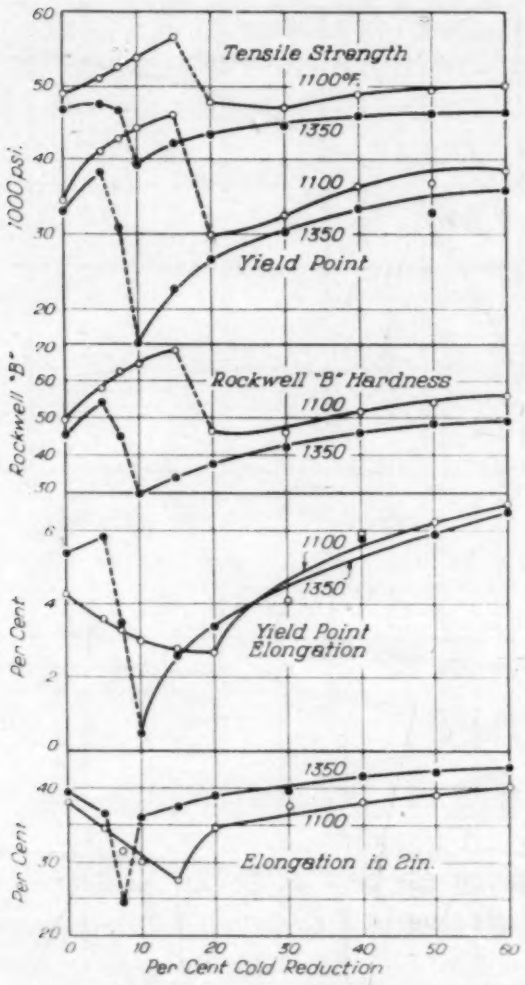

Fig. 12—Tensile Properties of Mild Steel Sheets Cold Reduced Various Amounts and Box Annealed at 1100 and 1350 Degrees Fahr.

3. For the 1100 degrees Fahr. anneal, Rockwell hardness, yield point, and tensile strength have a tendency to increase with percentage of cold reduction, up to the critical strain region. At 1350 degrees Fahr. this effect is not pronounced, but seems to be present.

Fig. 12 shows the effects of box annealing on Rockwell hardness, yield point, tensile strength, elongation in 2 inches, and yield point elongation, after annealing, and needs no further discussion.

Finishing

In the broad outline for this symposium, Section 4-d is concerned with Cold Straightening. It involves some of the processes

and their metallurgical effects, which generally are included in a group of operations called "Finishing." These processes involve temper rolling, roller leveling, and stretcher leveling.

One of the most fundamental sources of information which is broadly applicable to the metallurgical and flattening effects obtained from stretcher leveling and from so-called temper rolling and roller leveling, is the stress-strain, or the load deformation diagram obtained from tensile testing.

For discussion, a diagram of this kind is shown in Fig. 13, as obtained from fully-annealed material. It can be divided into the following portions:

1. Elastic deformation.
2. Yield point elongation.
3. General elongation.
4. Localized reduction of area.

The amount of elastic deformation up to the yield point for dead-soft mild steel sheets, is of the order of 0.1 per cent. In the process of flattening by stretching beyond the yield point, variations in the elastic recovery after release of load are the major causes for failure to obtain flatness. As will be readily appreciated, the variation in elastic recovery after releasing the loading in stretching, is due to the variation in unit stress existing in different parts of the stock subjected to the flattening process by stretching.

A more complete discussion of the relation of elastic behavior to the success of the flattening process by stretcher leveling, will be given in connection with the more detailed discussion of leveling equipment and its use.

Division 2 (Fig. 13) of the stress-strain diagram is designated as the yield point elongation region. It constitutes the wavy portion of the diagram immediately following elastic deformation. The yield point elongation portion of the stress-strain diagram is one of great practical importance, both in forming operations and operations used in flattening. It is also of very great metallurgical interest due to the considerable obscurity of the mechanism by which it results.

The following viewpoint makes possible a certain degree of understanding of the cause for this portion of the curve. In the stretching of a tensile test bar of sheet material within the plastic deformation region, there are two opposing forces in action: One is that of strengthening of the material due to work hardening; and the other that of decreased strength caused by reduction of area for those

portions being plastically deformed. If the rise in stress necessary to continue elongation for any portion of the test piece which is yielding for a sample of constant cross section is greater than the reduction of stress due to the weakening effect, then generalized elongation will result. This is the relation between these two effects which exists

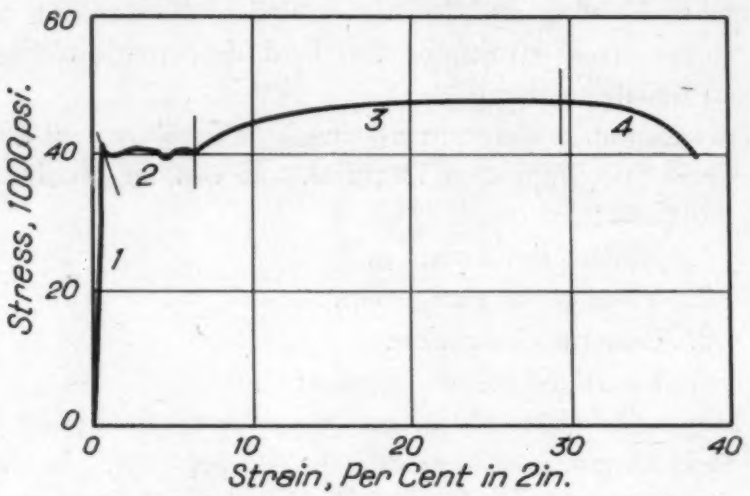

Fig. 13—Typical Stress-Strain Diagram of a Dead-Soft Steel Sheet.

in portion 3 of the diagram. If, however, the total weakening effect due to the reduction of area up to any given percentage of elongation is greater than the total rise in stress due to the work strengthening effect, then localized deformation results.

If the relation just mentioned continues long enough, the bar will pull in two at the first point at which deformation occurs. This is what happens in pulling tensile test bars of lead.

In mild steel, the piece does not pull in two at the point of first plastic deformation. Reduction of area at such a point proceeds to a certain amount, usually from 3 to 7 per cent, at which time plastic deformation starts at other points.

In actual testing of dead-soft mild steel test bars, usually the first plastic deformation occurs near the shoulders—after which deformation progresses from these two points toward the middle section of the test bar. The reason for the initiation of deformation at these places is that stress concentration occurs here, because of the rather sharp variation in width of section. The reason for the continuous progress of plastic deformation from the region near the shoulders toward the center is that concentration of stresses occurs at these points where rather abrupt changes in thickness of section has taken place.

It is obvious that if localized deformation has reduced by, say 6 per cent, the cross sectional area of the test piece at one point, a very considerable amount of work strengthening must occur in this reduced section if the stress value rises high enough to start plastic deformation in the unchanged area. This behavior has some important consequences:

1. If elongation by tension is stopped before all portions of the metal are reduced enough, inequalities in thickness of section and in surface level remain. This condition is usually known by stamping producers as stretcher strains or "worms."
2. In the process of making a stamping, a wide range of elongation exists for different portions of the stamping. Consequently, stretcher straining occurs if any considerable elongation at the yield point is present as the steel goes into the press.
3. If the piece is elongated by an amount equal to, or greater than, the yield point elongation, all parts of the test section are essentially equal and the surface has to a considerable extent returned to a smooth condition. Stretching through the yield point elongation region is usually known as "pulling through the stretcher strains."
4. In the manufacture of stampings, steel sheets or strip must have the yield point elongation removed if satisfactory surface is to be obtained.
5. Strain aging may cause yield point elongation to recur after tempering.

In commercial cold-rolled steel fully annealed, the yield point elongation may range from about 3 per cent up to as much as 50 per cent of the total elongation. Usually in sheets or strip of acceptable drawing properties untreated for stretcher strain, the yield point elongation will be within the range of from about 3-7 per cent. Some of the factors with which yield point elongation is related are:

Yield point elongation is related to elastic ratio, $\frac{\text{yield point}}{\text{tensile strength}}$. The lower the elastic ratio, the less the yield point elongation.

Elongation at the yield point is usually greater on fine-grained material than on coarser grained material. (See Figs. 8 and 11.)

In Fig. 12 is shown the variation in yield point elongation with various percentages of cold reduction before box annealing at two different temperatures. The variation in yield point elongation, as

shown in this figure, is related to variation in grain size, which in turn depends on the amounts of cold reduction used before annealing.

Consideration of the plastic deforming stresses as affected by work strengthening on one hand, and by reduction of cross sectional area on the other, as already stated, indicates that elongation at the yield point is due to a low rate of work strengthening during early stages of plastic deformation. Among a number of scientific workers, there is the view that two different types of structural changes are in progress during plastic deformation, one which has been termed "Fragmentation," and the other "Deformation by Slip." According to this view, one may assume that fragmentation, so far as it affects the breaking-up of some structure, might constitute a weakening effect, or at least it might result in a low rate of work strengthening.

The work strengthening effect of slip plane deformation may be due to the bending of planes of easiest slip, together with the development of preferred orientation. Part of the strengthening effect of the development of preferred orientation no doubt is due to planes of slip taking positions with respect to the distorting forces, which offer greater resistance to further deformation. If the fragmenting process is more predominant during initial stages of deformation than is the development of orientation, lower rates of work strengthening may result.

Further consideration regarding true stress curves showing a continuous rise in the stress with deformation, even for metals which show elongation at the yield point, leads us to believe that data are not available because of the difficulty encountered in measuring such highly localized deformations.

Proceeding with the discussion of the two remaining portions of the stress-strain diagram, it may be said that the smooth stress rising portion of this curve is obtained when the rate of work strengthening is greater than of weakening. The fourth portion of this curve is characterized by a reduction in load as elongation proceeds. It may be accounted for by the fact that the rate of work strengthening has fallen to values that are too low to make up for the weakening effect caused by reduction in cross section. This rapid action of necking-down at one point just precedes failure.

Elimination of Yield Point Elongation (Preventing Stretcher Straining)

The method used to prevent stretcher strain consists of subject-

ing the material to small cold reductions (from 0.5 to 2 per cent) by rolling on a "temper mill." In Fig. 14 is shown a photograph of a series of temper mills. The elimination of stretcher strain by this method consists in effecting reduction by sufficient cold rolling to eliminate the yield point elongation.

The relation between the yield point elongation and the amount of temper rolling required are shown in Fig. 15. The following are

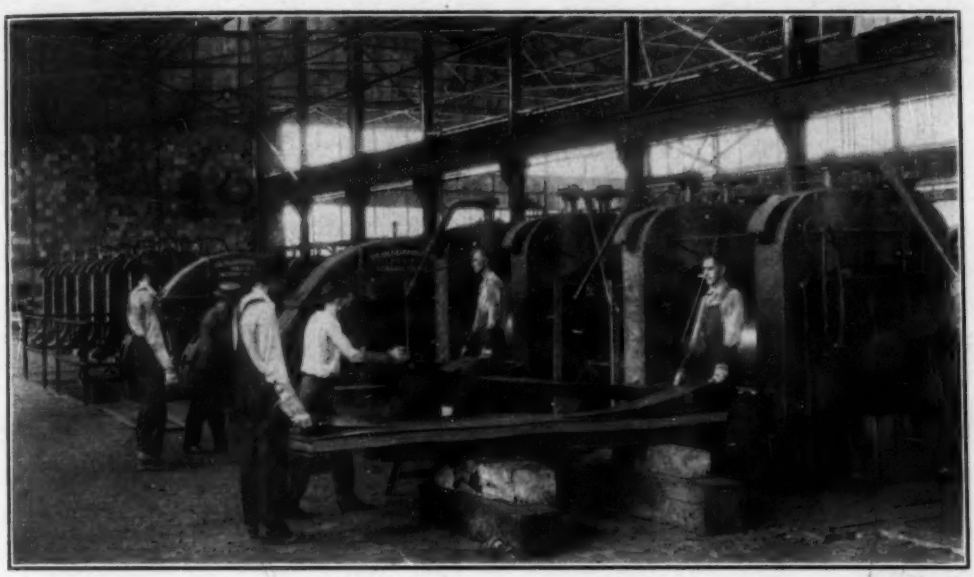

Fig. 14—Temper Rolling Mills.

the more important of our observations in regard to effects of types of temper mills and metallurgical characteristics of stock on the amount of temper rolling necessary to eliminate strain:

1. The smaller the grain size of the steel, the greater the yield point elongation; and, the greater the amount of temper rolling necessary.
2. Sheets with polished or smooth surface with the same yield point elongation, require more temper rolling than do sheets with dull surfaces.
3. The smaller the rolls for tempering, the smaller the amount of reduction necessary. (See Fig. 16).

Flattening or Leveling

Fig. 17 shows three roller levelers. The roller levelers commonly in use have a number of top and bottom rolls. The upper and lower rolls are staggered so that in passing through the leveler, the sheets are flexed alternately up and down. This has the effect of cold

working the surfaces more than the interior of the sheet. With levelers of the proper amounts of staggering of rolls, flatness may be obtained with essentially no increase in length of the sheet.

At the entrance side of the leveler, the axes of top and bottom

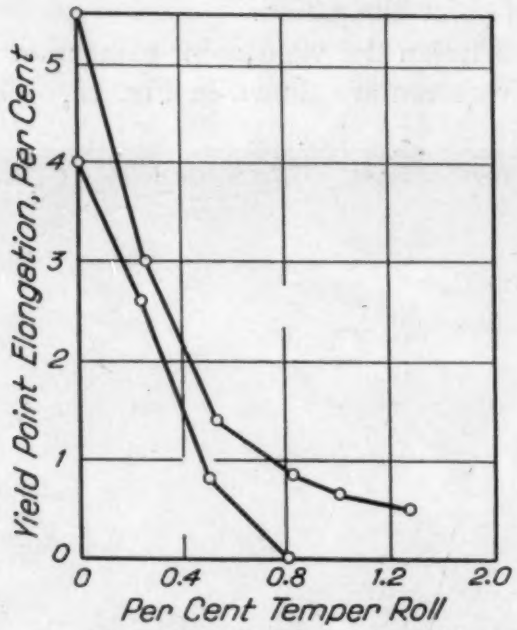

Fig. 15—The Effect of the Amount of Yield Point Elongation as Annealed on the Amount of Temper Rolling Necessary to Eliminate the Yield Point Elongation. Temper Rolling Done with 4-Inch Diameter Rolls. Upper Curve Sample A, Lower Curve Sample B.

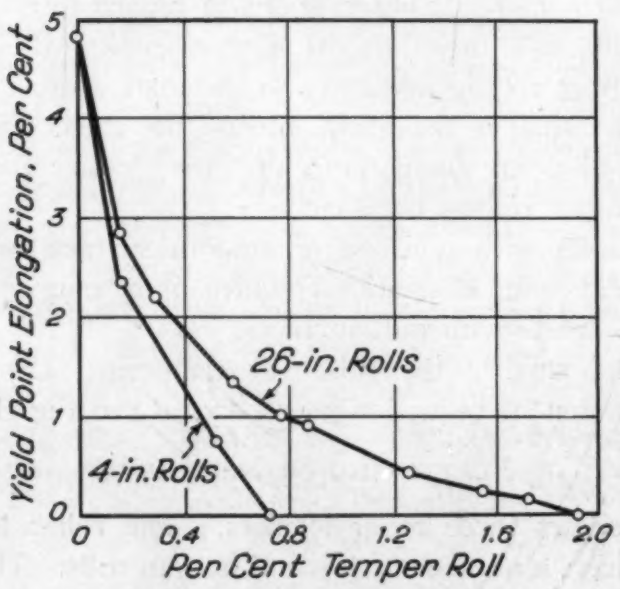

Fig. 16—The Effect of Roll Diameter on the Amount of Temper Rolling Necessary to Remove the Yield Point Elongation.

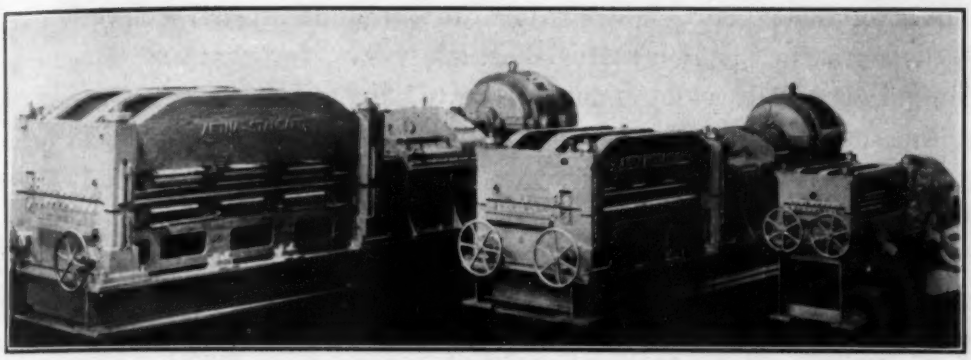

Fig. 17—Roller Levelers.

rolls are closer together than at the exit side. This results in a heavier flexing in passing through the earlier rolls of the machine. At the exit side the action is chiefly that of straightening. Prior to leveling, it is the usual practice to run the material through a temper mill to remove most of the yield point elongation. Unless this is done, breaks in the stock occur near the entrance side of the leveler.

A photograph of a stretcher leveler is shown in Fig. 18. This method of flattening is resorted to when extreme requirements in flatness must be met. As in the case of the roller leveler, it is necessary to have treated the stock by temper rolling before subjecting the material to the stretcher leveling operation.

There are some interesting facts which are worthy of mention in connection with the elimination of strain by cold rolling on the one hand, and by stretching on the other. Application of the analysis already given of the stress-strain diagram indicates that stretching must be carried to a point which takes up all of the elongation at the yield point, if stretcher straining is to be avoided. This would mean that stretcher leveling must be carried through elongations from about 3 per cent, as a minimum, to some 8 per cent. It should be stated

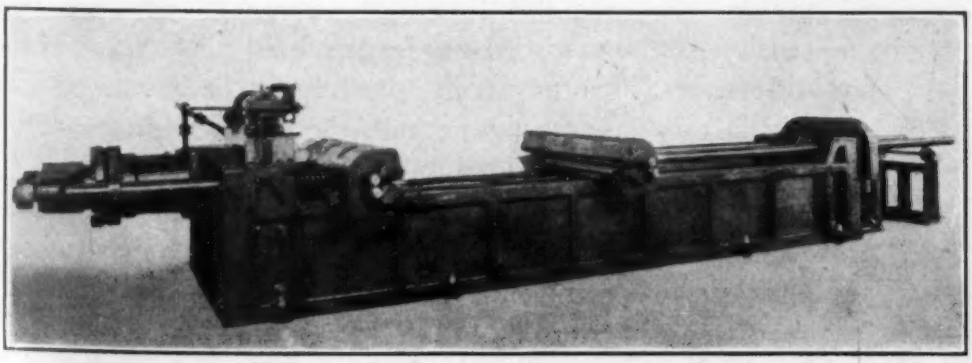

Fig. 18—A Stretcher Leveler.

that stretcher leveling is never used to eliminate strain; it is only used to obtain flatness. If a material which shows, for example, 4 per cent elongation at the yield point in the tensile test is elongated by cold rolling as little as 0.80 per cent, all of the elongation at the yield point will have disappeared. In accordance with what has already been stated, tensional stretching of this same material must have been carried to at least 4 per cent in order to have eliminated yield point elongation. This wide difference in values by cold rolling and tension is due to the fact that in the tempered material it is not necessary for the work strengthening process to make up the weakening effect of the reduction of section, which would have resulted in case yield point elongation were removed by stretching only. Stated in another way: It is only necessary to plastically deform the material by cold rolling to a point where the rate of work strengthening has risen to a value greater than that of weakening, due to reduction in cross section. The failure to obtain stretcher strains in low percentages of cold rolling is due to the fact that in the cold rolling operation, all parts of the material are reduced to the same cross sectional area as they pass through the rolls.

Experience shows that there are certain other requirements in sheet and strip products as they come to the stretcher leveling operation if flatness is to be obtained. In general, only certain types of nonflatness can be corrected by stretcher leveling. Some materials which are distinctly more out-of-flat than others, can be flattened satisfactorily by stretching.

A few typical cases will be mentioned:

1. Material which has a variation in physical properties across the sheet usually does not stretcher level satisfactorily. The difference in physical properties at various points across the sheet produces changing values for unit tension if the grips on the leveler hold perfectly. These differences in unit stress produce variation in strain, within the elastic range, across the sheet, with the result that release of the load produces different amounts of elastic contraction which causes nonflatness.
2. Nonuniformity in gage across a sheet of material whose physical properties are uniform in this direction, frequently gives trouble in stretcher leveling due to difficulties in obtaining perfect gripping. Slippage in the grips

will produce different unit stresses in different parts of the sheet. If this slippage is high enough so that stretching is not carried definitely beyond the yield point in all parts of the material, varying amounts of elastic contraction will result upon release of load, which in turn produces nonflatness.

Then, too, slippage causes different amounts of elongation in various parts of the sheet. This produces unsatisfactory flatness.

Great improvements in stretcher levelers have been made in the last few years. Instead of building the grips all in one rigid piece, they are made in short sections. This construction makes possible a more satisfactory adaptation of grips to slight variations in gage.

3. As the physical properties of sheet material rise, the leveling operation becomes more difficult because of the greater elastic contraction upon release of load, as well as greater difficulties in obtaining perfect gripping in the leveler.

Experienced operators of stretcher levelers have come to recognize certain conditions of nonflatness and of physical properties which can, or cannot, be successfully leveled. When shapes of stock, physical properties, and various degrees of nonflatness are considered, it is easy to appreciate how very complex problems in stretcher leveling may arise.

The discussion of many of these special features is beyond the scope of the present paper.

Acknowledgments

The authors take this opportunity to express their indebtedness to Messrs. R. J. Raudebaugh and R. L. Solter, of the Research Laboratories, for their co-operation in preparing the data and curves. Also, to the Aetna Standard Co., for the use of photographs of a stretcher leveler and roller levelers. Certain portions of the theoretical treatment on the mechanism of cold reduction were obtained from references which have been lost.

DISCUSSION

Written Discussion: By Joseph Winlock, chief metallurgist, Edward G. Budd Mfg. Co., Philadelphia.

I have read the paper by Dr. Hayes and Mr. Burns with a great deal of pleasure and I should like to congratulate them for their constructive work. Their paper contains a great fund of information which will be of great value to those interested in the manufacture of steel for deep drawing. It should be of particular interest, also, to those engaged in the use of steel for deep drawing, i.e., to the men in the press shop. It cannot but help to lead the way to a better co-operation and understanding. Too often, I think, those engaged in one industry fail to appreciate the problems which confront those whose product they are using. This often leads to erroneous conclusions and misunderstandings which seriously impede real progress. This paper clearly states some of these problems and the constructive work which has been done on them.

The curve in Fig. 15 shows clearly that if the elongation at the yield point is over 5.0 per cent, a relatively large reduction by cold rolling is necessary to prevent the occurrence of "stretcher strains" in the finished stamping. It has been our experience that yield point elongations of over 5.0 per cent are not unusual with steels of small grain size such as are frequently encountered in cold reduced strip which has been box annealed only.

Fig. 16 shows the extremely important finding that stretcher strains can be prevented from occurring in steel for deep drawing by temper rolling on small diameter rolls with much less reduction than by temper rolling on large diameter rolls. Dr. Hayes and Mr. Burns' figures show that 150 per cent more reduction is necessary to cause the removal of the yield point elongation (stretcher strains) when large rolls are used rather than when small rolls are employed. I should like to ask the authors if they can give an explanation of this interesting phenomenon.

Both of these curves are of particular interest and, it seems to the present writer, of great importance in obtaining a steel of maximum ductility after temper rolling.

Written Discussion: By E. A. Matteson, Aetna-Standard Engineering Co., Youngstown, O.

May I take this opportunity to congratulate Dr. Hayes and Mr. Burns on their very excellent and complete paper on this subject.

In passing over the discussion of the various methods of cold reducing, I might add that recent installation of feed reels and pay-off reel equipment have been arranged for regenerative braking, which gives a much better control over back tension and does away with the excessive heat and wear produced when a hand brake is used. With this method, a considerable portion of power loss is returned to the line.

The mention of elastic recovery is extremely interesting, as this fact gives us a better understanding of some of the reasons why some sheets will flatten satisfactorily and others will not, regardless of method used.

Leveling is still an art rather than a science, but it is probable that, with a better understanding of some of the peculiarities of cold-rolled steel, leveling can be brought closer to a science. A study of Figs. 9 and 13 reveals the reason for breaks or worm lines when dead soft steel is roller leveled, and their absence after temper rolling.

As the sheets are necessarily strained to the elastic limit when roller leveled and due to the practical ratio between the roll diameter and thickness

of material, it cannot be stressed much beyond this point, and therefore falls nicely within the range of elastic elongation. As it is possible to stretch the steel through this zone of elastic elongation, it should be reasonable to assume, that if it were possible to use rolls small enough in diameter, it would be possible to roller level dead soft material.

During a recent experiment in leveling annealed cold strip 0.037-inch thick with a leveler having 1¾-inch rolls, it was found that nearly 0.3 per cent elongation was obtainable and at this elongation strain lines or breaks which were bad with a light pass and little elongation became fainter. It is not my intention to suggest that temper rolling might possibly be done away with, for it is doubtful that sheets worked this hard in a leveler would have good physical characteristics for deep drawing. However, with the use of small roll levelers, the tendency to stretcher strain can be removed from sheets which have been skin rolled. This tendency to stretcher strain often remains when sufficient skin rolling to remove them is not allowable due to certain limitations in hardness specifications.

A sheet, to level properly, should be slightly dished. The smaller the leveler rolls, the greater will be the amount of permissible dish, and still obtain flatness.

As it is much easier to control a skin mill with relatively heavy crown, it follows that the smaller the leveler rolls used, the easier becomes the temper rolling and the greater the percentage of stock which will be commercially flat.

Very little can be added to the discussion of stretcher leveling which covers general practice very well. Recently, however, a demand for good drawing sheets with stretcher leveled flatness, has produced a slightly new method. It is well known that when a sheet is stretched to the 1 to 1¼ per cent necessary for flatness, it has lost much of its drawing quality. To overcome this, the sheet, when given a severe roller leveler pass, will stretcher level with practically no elongation, that is by simply pulling taut without hydraulic pressure, and will satisfactorily meet the above mentioned demand. The two operations must be consecutive for best results.

Written Discussion: By Dr. Charles W. MacGregor, Instructor of Mechanical Engineering, Massachusetts Institute of Technology, Cambridge, Massachusetts.

The practical importance of the "yield point elongation" has been brought out clearly by the authors. The relations between "the yield point elongation" and the per cent temper rolling necessary to eliminate this elongation at the yield point as given in Figs. 15 and 16 should prove of considerable use.

It has been shown in the paper that roll size affects the necessary amount of temper rolling. Various mild steels, however, have different modes of transition from the elastic to the plastic state[1] and the formation of stretcher strains depends on the sharpness of this transition. The writer would be interested to know, therefore, if this mode of transition has any effect on the problem of selecting the correct amount of temper rolling. In other words,

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[1]C. W. MacGregor, "The Yield Point of Mild Steel," *Transactions*, American Society of Mechanical Engineers, APM 53-5, p. 187-200, 1931.

will two steels of the same total yield point extension but of different modes of transition require the same amount of temper rolling?

The question of the fundamental mechanism of plastic flow taking place during the cold rolling of a steel bar or strip is indeed both an interesting and a complicated problem. Certain valuable tests have been made in Germany by Siebel and Lueg[2] in which the contact pressure distribution between the roll and the material passing through the roll has been measured by piezo-crystals. This should provide an important step in the analysis of the stress distribution in the plastic state in the material as it passes through the rolls. Sometime ago, the writer found that the pressure distribution between a bakelite strip, cut in the shape of a rolled bar, and two bakelite rolls compressing the strip was approximately similar to that found on the actual cold rolling tests by Siebel and Lueg. It is of course realized that a photoelastic test made on bakelite specimens does not reproduce many of the conditions present in the rolling problem. The similarity of pressure distribution, however, is worthy of note. This is of course a very complicated problem, and can only be attacked more successfully than it has been until now after further experimental information on such questions is made available.

Oral Discussion

H. P. MUNGER:[3] The authors have presented interesting data on the production of cold reduced mild steel sheets and strip. It is of interest to carry their work on the advantages of front and back tension one step farther and apply it to a tandem mill. They have pointed out that only when the energy liberated in producing back tension is utilized, is it economical to take advantage of the decreased screw pressure and the decreased power required for reduction in gage. Consider for example a three stand tandem cold mill. For economical operation, the entry stand should have a back tension reel coupled to an electric generator. The second stand may have back tension which, in turn, will produce front tension on the first stand. Similarly, tensions between the second and third stands may produce front and back tension on these stands, respectively. A power reel may apply the front tension to the last stand. In this way, the decrease in roll pressure due to back tension and the decrease in power required for the reduction to gage may be utilized. It is understood that some of the cold mills are using such a method of rolling.

The elastic ratio, that is, Yield Strength divided by Tensile Strength, which the authors have mentioned in connection with the yield point elongation, is especially interesting in the case of deep drawing steels. It may be considered as the drawing range in which the steel sheets may be drawn. It is necessary to exceed the yield strength to produce plastic deformation, and when the tensile strength is exceeded, localized necking-down produces breakage, hence the term drawing range. It is obvious that the drawing range is bounded by the yield strength and the tensile strength, and may be represented by the elastic ratio.

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[2]E. Siebel und W. Lueg, Mitt. aus dem Kaiser Wilhelm Institut für Eisenforschung, Vol. 15, p. 1, 1933.

[3]Republic Steel Corporation, Warren, Ohio.

The elastic ratio under uniform testing conditions may be useful also in determining the probable annealing treatment of low carbon steel sheet. For normalized steel a value of about 0.75 is frequently obtained. If this sheet be box annealed, this value will usually drop to the neighborhood of 0.65. On temper rolling this sheet, say 1 per cent, the ratio may drop to 0.55 or 0.60. The latter decrease in elastic ratio is due to a decrease in the yield strength, and a small increase in the tensile strength. In accordance with the authors' suggestions, this decrease in yield strength may be due to a continuation of the fragmentation process started by temper rolling, or it may be due to the increase of the number of nuclei from which plastic deformation will start, as Winlock and Leiter will point out in the next paper. If a large number of nuclei are present, the threshold value necessary for the initiation of plastic deformation, which is commonly called the yield strength, will be lowered.

With reference to stretcher leveling, it has been pointed out that sheets which are not homogeneous in cross section, i.e., have been straight rolled, and include a segregated area with unsegregated skin at either edge, are not suitable for this process. This uniformity of properties across the width of the sheet may be obtained either by cross rolling or by using killed steels.

Authors' Closure

We appreciate Mr. Winlock's kind remarks and realize that those engaged in one industry sometimes fail to understand the problems which confront those whose product they are using. Mr. Winlock asks for an explanation as to why smaller rolls are more effective in producing freedom from stretcher strains. J. S. Caswell, in his article in the *Engineer,* February 28, 1930, shows that with the same draft the amount of stock undergoing compression simultaneously is greater with the larger roll. The arc of contact is greater and therefore the friction zones are larger and they penetrate deeper into the stock. These zones exert a greater constraint on the movement of the middle zones and the difference between the stress conditions after the pass in the outer zones and the middle zones of the stock will not be as great as that associated with the smaller roll. This means that materials temper rolled on smaller rolls are in a more unstable condition with respect to internal stresses near the surface and near the mid-thickness of the sheet. The fact that both temper rolling and quenching produce a decrease in the amount of elongation at the yield point suggests a possible interesting parallel. At present insufficient reliable data are available to make a satisfactory explanation.

Dr. MacGregor refers to different modes of transition from the elastic to the plastic state. We believe that so long as the yield point elongation is a constant in a given set of samples, that the mode of transition from the elastic to the plastic state depends more on such things as sample shape, testing speed, and the elasticity of the testing machine used.

Dr. Munger, in giving the elastic ratios for normalized and normalized, second annealed sheets, states that the elastic ratio drops from 0.75 to 0.65 in box annealing. We believe that the elastic ratio drops very little on second annealing because the yield strength and tensile strength are both decreased

on box annealing after normalizing. However, this paper deals only with cold reduced box annealed sheets, the elastic ratio of which is about 0.60 to 0.68 as box annealed, and 0.55 to 0.63 as temper rolled 1 per cent. It is hard to understand how the number of nuclei from which plastic deformation starts can have a bearing on the yield strength value. The number of nuclei from which plastic deformation starts does, however, have a bearing on the yield point elongation, as shown by the work of Winlock and Leiter.

Both Dr. Munger and Mr. Matteson point out methods of retrieving the power loss at the break if high back tension is to be applied. In this paper we merely mean to point out the differences in power required by the use of high and low back tension.

SOME FACTORS AFFECTING THE PLASTIC DEFORMATION OF SHEET AND STRIP STEEL AND THEIR RELATION TO THE DEEP DRAWING PROPERTIES

By Joseph Winlock and Ralph W. E. Leiter

Abstract

The ductility of sheet and strip steel for deep drawing purposes is substantially affected by the necessity of cold rolling the steel as a final operation in its manufacture in order to prevent the otherwise inherent occurrence of Luder's lines ("stretcher strains") during the deep drawing operation. In addition, such cold rolling increases the susceptibility of the steel to "age." If, however, the steel is subjected to exactly the proper amount of cold work immediately before the deep drawing operation, stretcher strains do not occur and the ductility is not appreciably affected. Some of the factors which influence these phenomena such as methods of steel manufacture, grain size, rate of deformation, etc., are described.

IN tracing the development of the industries engaged in the making and using of steel, it is not unusual to find that as the methods of manufacture are improved, the requirements for certain physical properties of the steel become increasingly more exacting. This is not surprising, but is in simple conformity with the general order of progress. In the effort to progress more rapidly, however, it sometimes happens that unequal attention and emphasis are placed on different phases of the development. The result of this is that considerable lapses of time often occur in which the net increase in progress is small. Experience has shown, for example, that to insure a constant or increasing rate of progress, careful investigation and research in the metallurgy of the steel is equally as important as the mechanical development of the various processes involved in its manufacture and use.

In our studies of the metallurgy of steel for deep drawing purposes, it has been found that progress towards obtaining the most

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ductile steel is hindered by the necessity of effecting a compromise between the action of two phenomena which produce diametrically opposite results on the physical properties of the steel. These two phenomena are (1) that the annealing necessary to impart the greatest ductility to the steel also puts the metal in a condition in which Luder's lines[1] (known in the shop as "stretcher strains") occur on the surfaces of the metal during the deep drawing operation, and (2) that the cold rolling of the steel as a final operation in its manufacture made necessary to prevent stretcher strains from occurring, lowers substantially its ductility. Cold rolling as a final operation also makes the steel more susceptible to "aging." As is well known, the result of aging is a spontaneous increase in hardness and a spontaneous decrease in ductility. This has been pointed out in some studies of steel for deep drawing by one of the present authors[2], by Griffis, Kenyon and Burns,[3] and by Davenport and Bain.[4]

Object of Experiments

The purpose of this paper is to discuss the progress which has been made in an attempt to throw some additional light upon some of the factors influencing the formation of Luder's lines. It is manifestly clear that such studies are extremely important because if their occurrence could be more accurately controlled, or if they could be prevented from occurring in fully annealed steel, the difficulty of obtaining and maintaining the maximum ductility in the material would be considerably lessened.

Luder's Lines

Luder's lines or stretcher strains appear on the surfaces of the steel after deep drawing as irregular lines of depressions or elevations which destroy the evenness of the surfaces of the metal (see Fig. 1). If the stresses producing the deformation are primarily tensile, the lines appear as depressions in the surfaces, but if the stresses are primarily compressive, irregular lines of elevations

[1]These "lines" are known also as Hartmann lines, the Piobert effect, etc.

[2]Winlock and Kelley, "Sheet Steel and Strip Steel for Automobile Bodies," Transactions, American Society for Steel Treating, Vol. 18, 1930, p. 254.

[3]Griffis, Kenyon and Burns, "The Aging of Mild Steel Sheets," Yearbook, American Iron and Steel Institute, 1933, p. 142.

[4]Davenport and Bain, "The Aging of Steel," Transactions, American Society for Metals, Vol. 23, 1935, p. 1047.

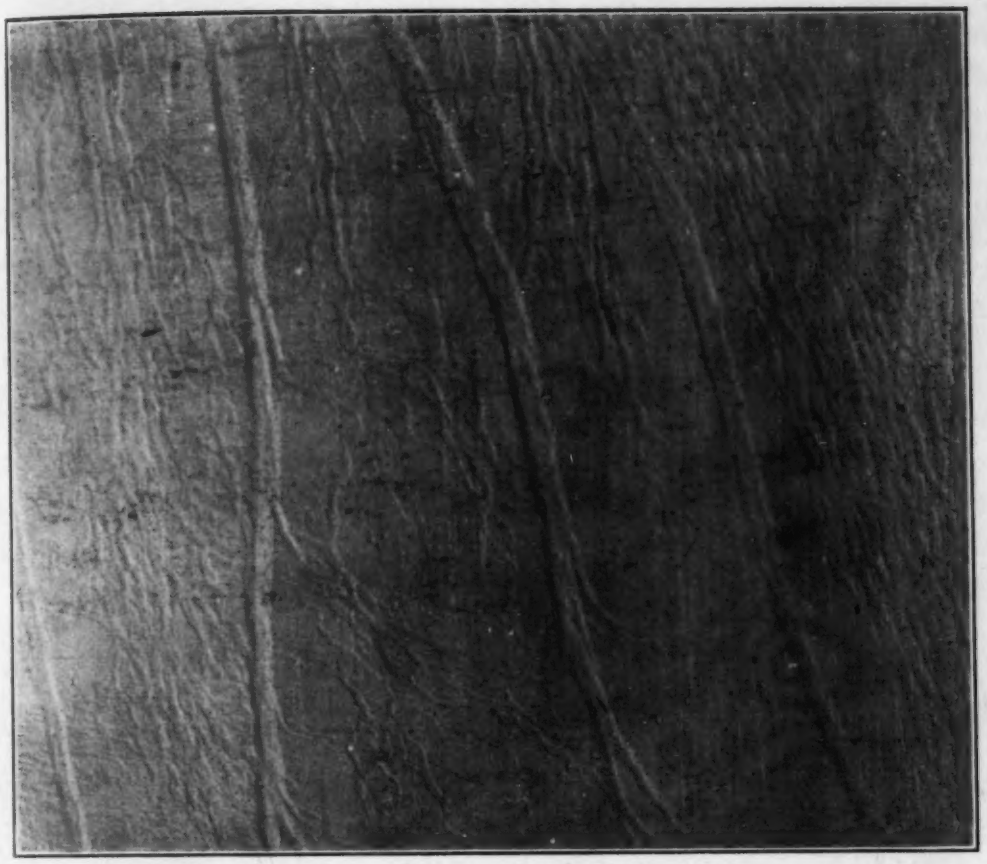

Fig. 1—Portion of Stamping Showing Stretcher Strains.

occur. It has been shown [5, 6, 7] that they occur as a result of the uneven and nonuniform flow of the metal at the yield point caused by the sudden transition of different grains or groups of grains from the elastic to the plastic state. The region in which they occur on the load-deformation curve corresponds to an elongation of from 1.5 to 10.0 per cent. Stretcher strains tend to occur, then, in those stampings or those parts of stampings where the percentage elongation falls within these limits. It has been shown, also, that by carefully regulating the amount of cold work done immediately before deep drawing, the load-deformation curve can, without perceptible loss of ductility, be made relatively smooth from origin to fracture (see Figs. 2 and 3). Steel in the latter condition will not show stretcher strains after deep drawing, and, because the deformation takes place more uniformly, the physical properties for deep drawing

[5] Winlock and Kelley ibid.

[6] Rawdon, Bureau of Standards Journal of Research, 1928, Vol. 1, p. 479.

[7] Kenyon and Burns, "Autographic Stress-strain Curves of Deep Drawing Sheets," TRANSACTIONS, American Society for Metals, Vol. 21, 1933, p. 109.

must be considered as having been improved. In order to preserve the condition of "overstrain" denoted by the load-deformation curve in Fig. 3, it is necessary to make certain that a lapse of time sufficient for "aging" is not permitted to take place before the deep drawing operation is performed. This is because concurrently with aging the load-deformation curve will revert to its original shape and stretcher strains will again appear during plastic deformation.

The occurrence of a sudden transition from the elastic to the

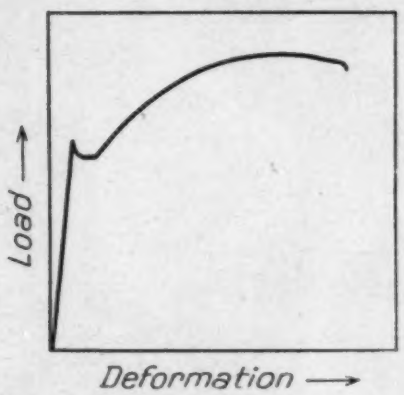

Fig. 2

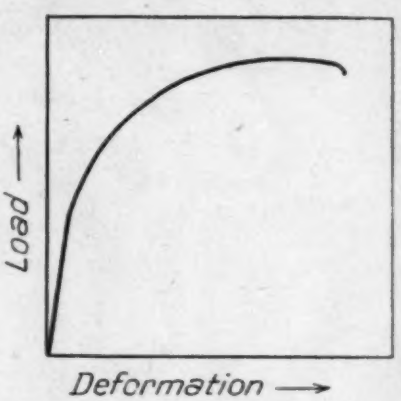

Fig. 3

Fig. 2—Load-Deformation Curve of Low Carbon Steel in Annealed Condition.
Fig. 3—Load-Deformation Curve of Low Carbon Steel After Cold Working.

plastic state in low carbon steel and the accompanying nonuniform and uneven plastic flow occurring at the yield point are most unusual and extraordinary phenomena. This peculiar property of iron and low carbon steel is not shown to any marked extent by most of the other metals in general use such as copper, brass, or aluminum. Several investigators have reported surface markings as having occurred during the plastic deformation of some of these metals notably Nadai for work-hardened copper,* Portevin and Le Chatelier for a tempered aluminum alloy,[8] and Sachs for crystals of beta brass.[9] There is considerable doubt, however, that the observed surface markings are of exactly the same character. At any rate, the present authors have never seen surface markings in the deep drawing of these metals which exactly resemble the appearance of stretcher strains.

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*"Plasticity," McGraw-Hill Book Co., p. 106, Fig. 95.

[8]*Comptes Rendus*, 1923, Vol. 176, p. 508.

[9]*Naturwiss*, 1928, Vol. 16, p. 412-416.

As is well-known, the deformation, occurring before the yield point of the steel is reached, appears by its behavior to be substantially elastic and the concomitant elongation appears to be uniform throughout the parallel section of the test piece. Within this range, the rate of strain in inches per inch per minute is substantially constant for any given head speed of the testing machine. When, however, the yield point is reached, the transition from the elastic to the plastic state does not occur throughout the test piece at the same instant, but at some local point or points. This means that parts of the specimen reach the yield point and begin to deform plastically before other parts. The stress at which such plastic deformation first occurs is, according to general testing practice, taken to be the yield point of the steel as a whole. It is the differences in thickness produced by the plastic flow occurring from point to point in the specimen which give rise to the surface markings known as Luder's lines or stretcher strains. During the time in which this irregular deformation is occurring, the load on the specimen remains practically constant. As this alternate local slipping and work-strengthening continues, and the extent and number of the depressions in the surfaces of the metal increase, the condition is reached where the entire test piece has been so affected. The differences in cross sectional area then become relatively minute. In effect, this is a return to the original even surface. The total elongation occurring during this peculiar process has been conveniently referred to as the "yield point elongation," and the depth of the stretcher strains has been shown to be proportional to this elongation. When all of the grains have slipped in this manner, the load-deformation curve begins to rise again as a relatively smooth curve.

Description of Testing Machine

In order to study more effectively the phenomena occurring at the yield point, a specially designed testing machine was constructed so that autographic load-deformation curves could be obtained at widely different rates of deformation. A photograph of this machine is shown in Fig. 4. The deformation is produced by a two-horse power induction motor (A), operating through a gear box (B) which revolves a screw (C), which, in turn, causes a wedge (D) to operate against the movable head of the machine (E). The specimen (F) is shown in the jaws of the machine (G and H). The

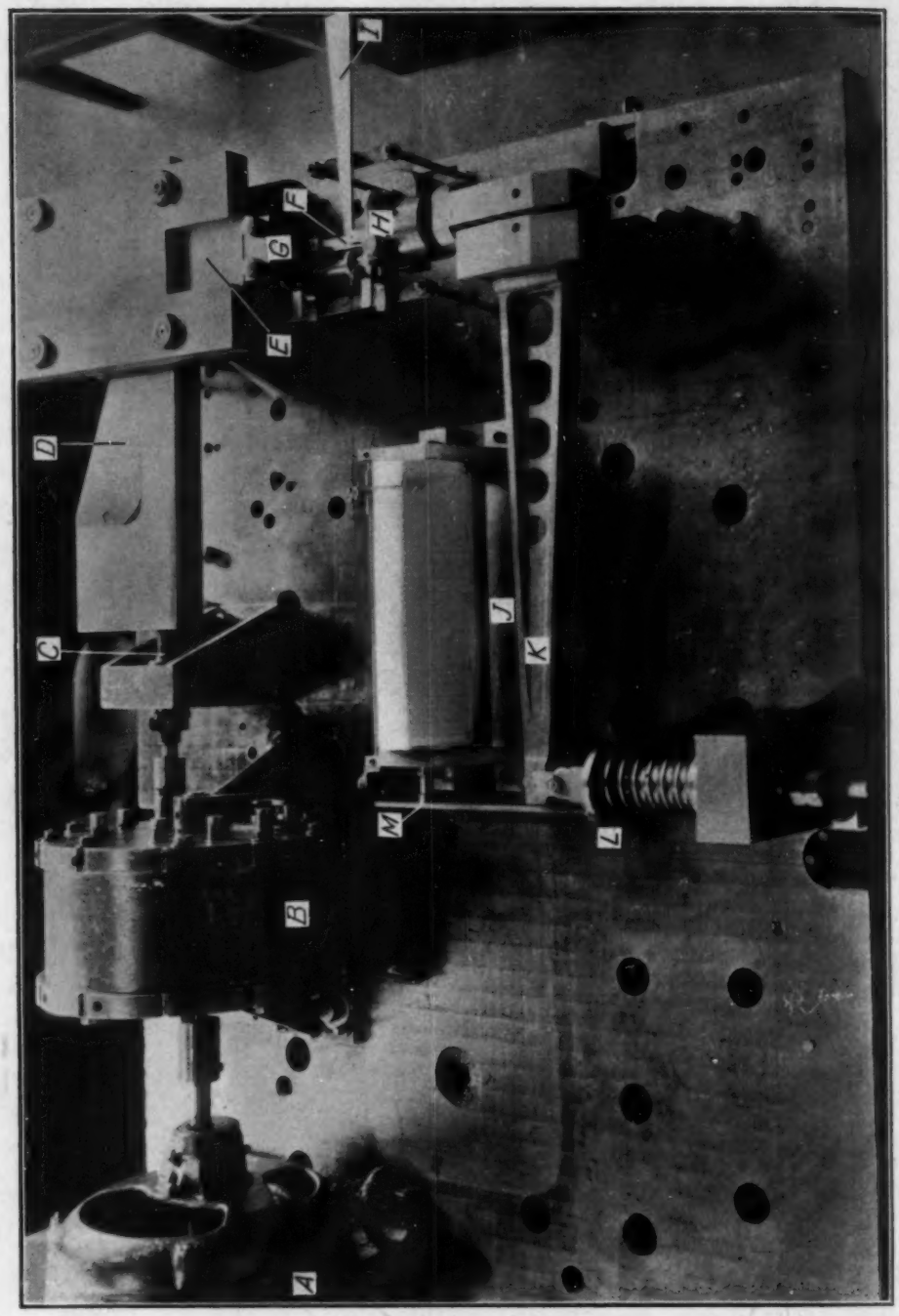

Fig. 4—Photograph of Testing Machine. See Text.

extension of the specimen is obtained by the wedge-type extensometer (I) and a record of this extension is obtained on the recording drum (J). The load on the specimen resulting from the deformation is transmitted by a beam (K) resting on knife edges to the calibrated spring (L). The movement of the spring is transmitted by the rack

Table I

Speed	Head Speed (Inches Per Minute)	Rate of Strain (Inches Per Inch Per Minute)
J	0.0046	0.002
I	0.0140	0.006
H	0.0417	0.018
G	0.125	0.056
F	0.375	0.166
E	1.130	0.50
D	3.380	1.50
C	10.100	4.44

and pinion (M) to the recording drum (J). The magnification of the extension was in all the tests 10 to 1. The beam has lever arms in the ratio of 16 to 1, and the magnification of the movement of the spring on the recording drum is in the ratio of 5.1 to 1. The maximum movement of the "fixed" head of the machine was in these tests less than $\frac{1}{16}$ of an inch. The gear box operating in conjunction with the wedge and screw was designed to give the eight different head speeds shown in Table I. The corresponding rates of strain (in inches per inch per minute) over the parallel section of the test piece are also given in this table. It is realized, of course, that the values for rate of strain are not exact because of the slight movement of the "fixed" head but for the purpose of these experiments as will be seen, the small differences in the actual rate are considered to be of no significance.

Steels Used

The steels used in these experiments were low carbon steels of the ordinary type used for deep drawing. In each case, the steel had been rolled to a finished gage of almost exactly 0.036 inch. The chemical analysis of each steel is given in Table II and, as may be noted, is practically the same in each case.

Steel number 1 was rolled by ordinary sheet mill methods, which include "cross" rolling, to approximately one-half the finished length, pickled, reheated and hot-rolled to the finished gage. This

Table II

	Steel 1	Steel 2	Steel 3	Steel 4	Steel 5
Carbon, per cent	0.05	0.05	0.05	0.06	0.04
Manganese, per cent	0.23	0.29	0.34	0.33	0.31
Phosphorus (less than), per cent	0.015	0.015	0.015	0.015	0.015
Sulphur, per cent	0.033	0.029	0.050	0.022	0.030

was followed by a "flattening" pass on the cold rolls and then by a box-anneal, at some 1250 degrees Fahr. (680 degrees Cent.) in a reducing atmosphere.

Steel number 2 was rolled by the same mill methods as number 1 except that the box-annealing was replaced by a "normalizing" treatment, at some 1700 degrees Fahr. (930 degrees Cent.). This was followed by pickling, a small amount of cold rolling, and a final box-anneal.

Steel number 3 was hot-rolled on a strip mill to approximately two and one-half times the finished gage and, after pickling, was cold-rolled to the finished gage. The steel was then normalized, pickled, given a "flattening" pass on the cold rolls and box-annealed.

Steel number 4 was rolled by the same mill methods as number 3 except that the reduction by cold rolling was followed by a box-anneal only.

Steel number 5 was made by the same mill operations as steel number 4.

An analysis of the various treatments will show that practically all of the ordinary mill processes commercially used today in the making of sheet and strip steel are represented in these five steels. The choice of the steels was also such that the effects of these different mill methods and treatments on stretcher strain formation, etc. could be noted in steels having the same and different grain sizes.

Tensile specimens were cut from the steels and machined to size. The specimens were in every case of the standard dimensions for thin sheet metals.[10] Such specimens have a gage length of 2 inches on a parallel section 0.5 inch wide and 2¼ inches long.

They were then heated in a sealed container to 1250 degrees Fahr. (680 degrees Cent.) and allowed to cool slowly in the furnace. The structures of the steels after this treatment at a magnification of 100 diameters are shown in Figs. 5, 6, 7, 8 and 9. These photomicrographs show the entire cross section of the sheet. This heat treatment was carried out in order to make certain that the effects of any cold work which may have occurred in handling or during the machining of the test specimens were first removed from the steels. It was carried out, also, to insure the removal of any of the effects of differences in the rates of cooling employed by the different steel mills. The removal of these variables was necessary because it is well recognized that, under certain conditions, changes in the

[10]American Society for Testing Materials Standards, Part 1, 1933, p. 954.

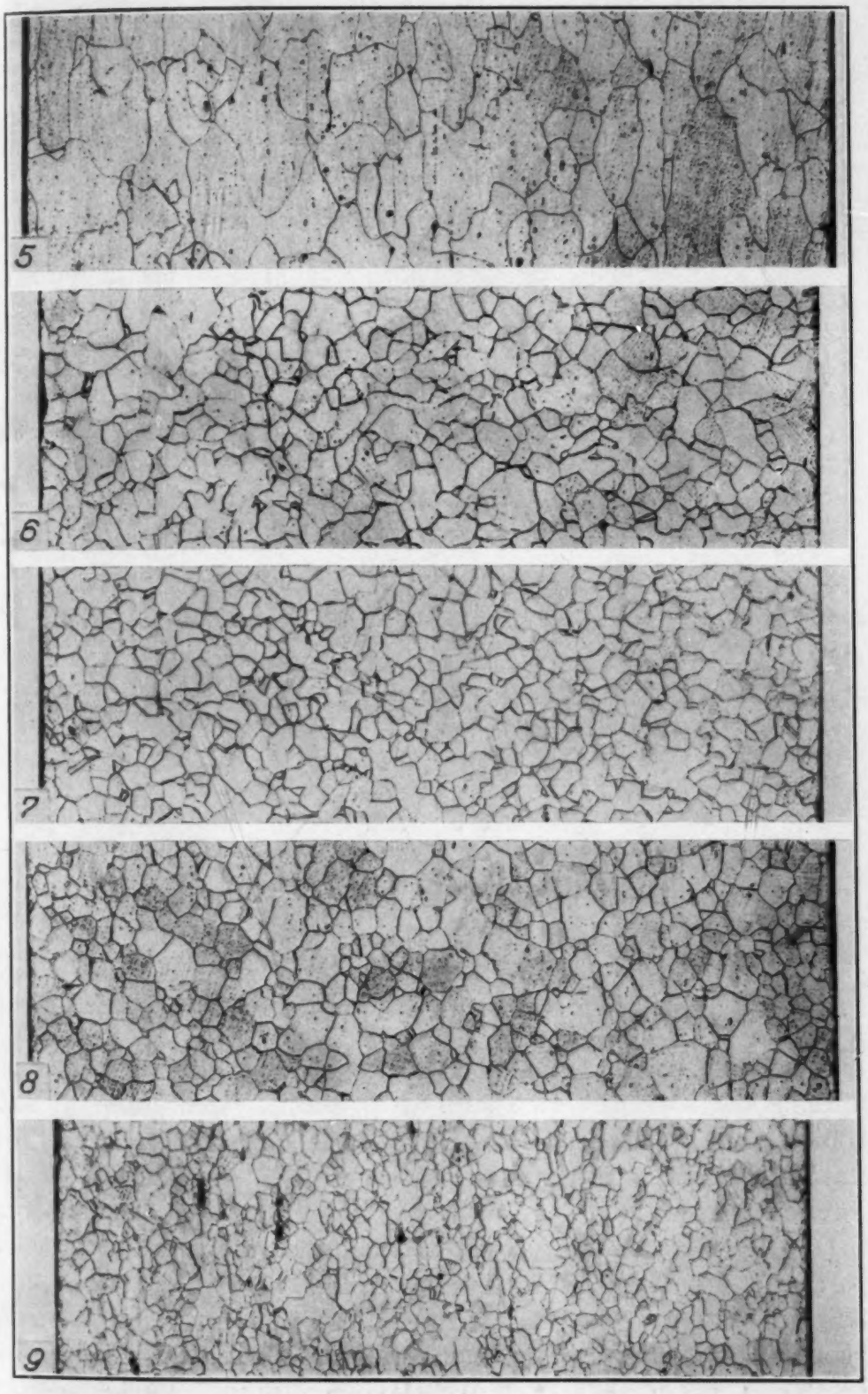

Fig. 5—Steel 1. 20 Grains Per Square Inch. (A. S. T. M. Grain Size 5.) × 100.
Fig. 6—Steel 2. 84 Grains Per Square Inch. (A. S. T. M. Grain Size 7.) × 100.
Fig. 7—Steel 3. 107 Grains Per Square Inch. (A. S. T. M. Grain Size 7.) × 100.
Fig. 8—Steel 4. 91 Grains Per Square Inch. (A. S. T. M. Grain Size 7.) × 100.
Fig. 9—Steel 5. 200 Grains Per Square Inch. (A. S. T. M. Grain Size 8.) × 100.

Table III

Steel No.	Thickness (Inches)	Speed	No. of Grains per sq. in. at 100 Diameters	Yield Point (Pounds per sq. inch)	Yield Point Elongation (Per Cent)	Total Elongation in 2 inches (Per Cent)	Tensile Strength (Pounds per sq. inch)
1	.0367	J	20	18,900	1.35	36.2	36,600
		I		19,800	1.50	35.5	37,000
		H		21,000	1.87	39.2	37,700
		G		23,000	2.4	38.8	38,400
		F		25,000	3.2	37.3	39,400
		E		27,300	3.6	38.5	39,300
		D		29,000	4.2	37.5	40,300
		C		30,400	4.4	38.9	41,400
2	.0365	J	84	26,500	2.1	38.5	42,800
		I		26,700	2.3	39.5	43,100
		H		28,600	2.8	41.7	43,800
		G		30,400	3.5	40.0	44,000
		F		33,200	4.3	40.7	45,200
		E		35,700	5.3	40.0	45,200
		D		38,600	7.2	37.5	45,700
		C		41,300	8.5	37.7	46,800
3	.0362	J	107	27,500	2.45	41.5	42,200
		I		27,600	2.65	41.3	42,700
		H		29,200	2.95	42.2	43,400
		G		31,400	3.7	44.2	44,100
		F		33,500	4.35	43.0	44,500
		E		36,800	5.6	40.5	45,100
		D		38,600	7.0	39.3	45,500
		C		41,800	9.4	41.1	46,400
4	.0375	J	91	24,600	2.4	46.5	39,100
		I		25,600	2.7	49.5	39,400
		H		27,100	3.2	48.0	39,500
		G		28,800	3.8	50.7	40,800
		F		31,100	4.8	45.0	41,400
		E		33,600	5.8	44.5	42,100
		D		35,600	7.5	43.0	41,600
		C		39,800	9.7	44.5	43,100
5	.0351	J	200	28,000	3.0	42.5	42,000
		I		29,800	3.6	...	.....
		H		31,400	4.35	47.5	42,800
		G		33,200	5.1	45.5	43,200
		F		35,600	6.2	44.3	43,500
		E		38,300	7.85	43.3	43,900
		D		40,300	10.3	40.5	44,700
		C		43,200	13.0	41.3	45,300

physical properties of the steel can take place without perceptibly altering the apparent grain size. For example, reductions by cold rolling even as large as 15 per cent cannot be detected under the microscope, but the physical properties are, nevertheless, deeply affected by reductions as low as 0.5 per cent. Again, changes in the aging characteristics occur as the rate of cooling from comparatively low temperatures is increased. This is because the solubility of carbon in alpha iron increases as the temperature is raised above room temperature. In general, the slower the cooling, the greater is the ductility of the steel from a deep drawing point of view.

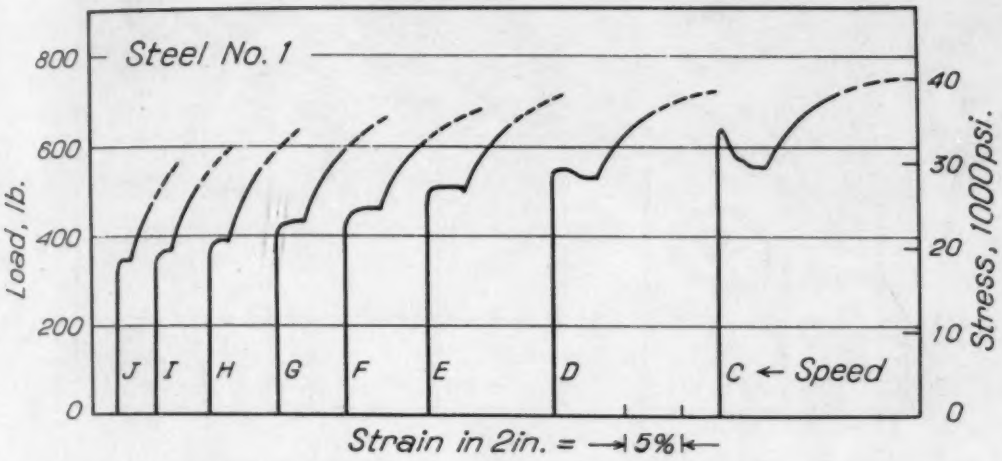

Fig. 10—Load-Deformation Curves at Different Constant Rates of Deformation. Steel No. 1.

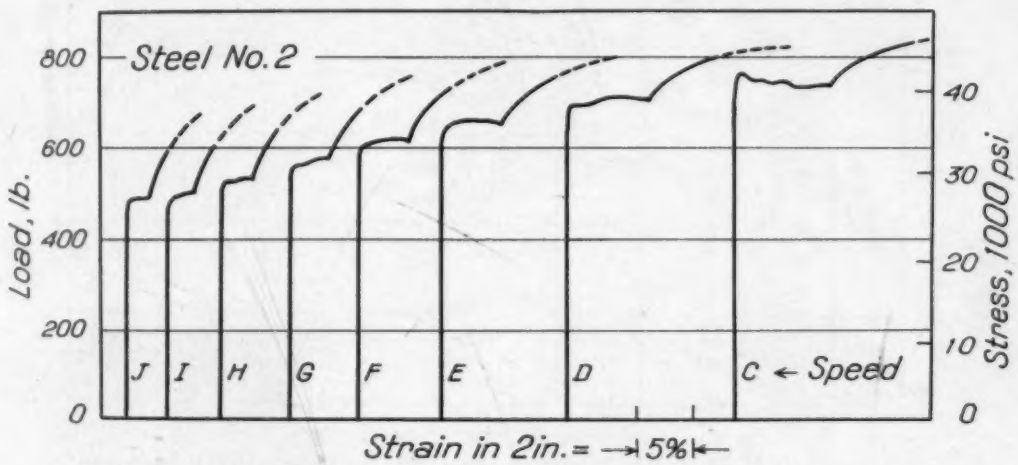

Fig. 11—Load-Deformation Curves at Different Constant Rates of Deformation. Steel No. 2.

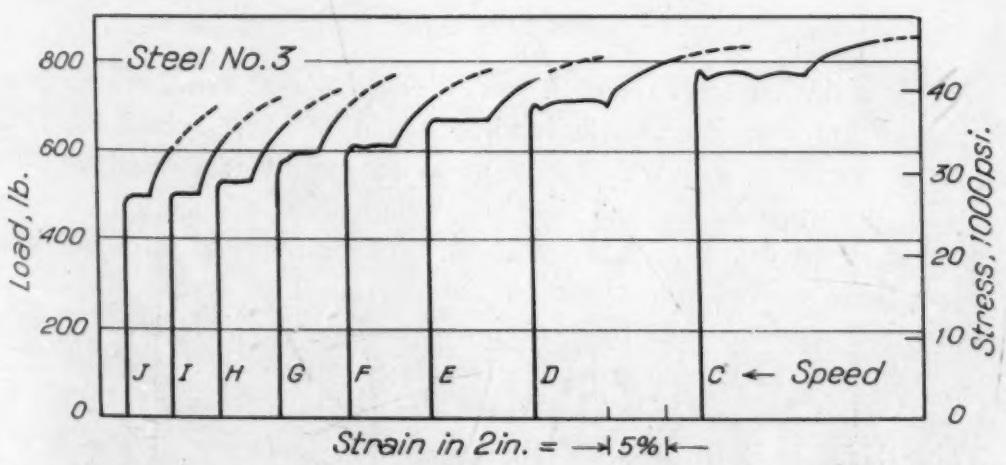

Fig. 12—Load-Deformation Curves at Different Constant Rates of Deformation. Steel No. 3.

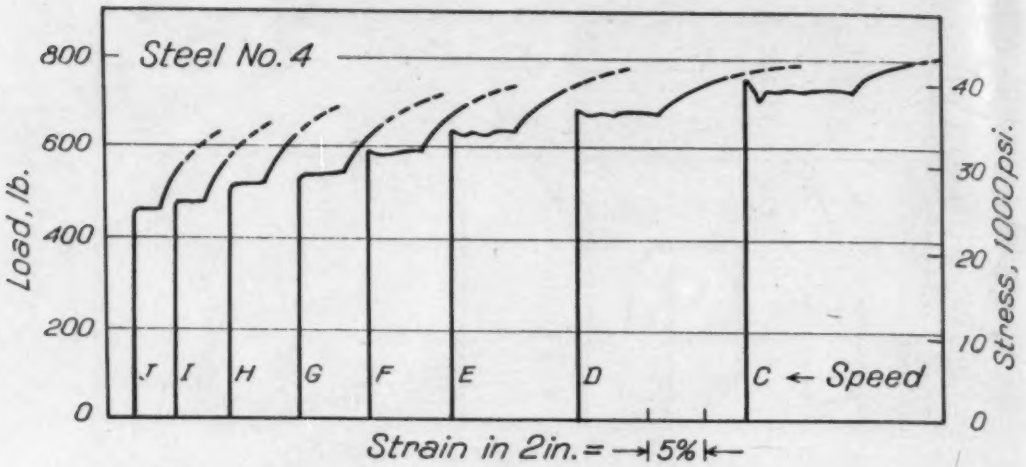

Fig. 13—Load-Deformation Curves at Different Constant Rates of Deformation. Steel No. 4.

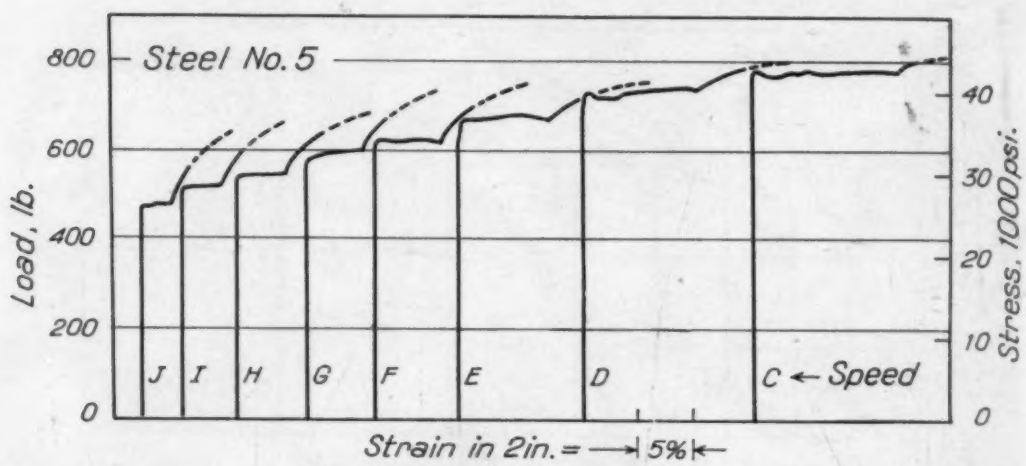

Fig. 14—Load-Deformation Curves at Different Constant Rates of Deformation. Steel No. 5.

Outline of Experiments

A series of tests on each steel were made at different constant rates of deformation and the load-deformation curves obtained. That part of the curve with which these notes are mainly concerned, viz., the elastic deformation and the yield point elongation, are shown for each steel in Figs. 10, 11, 12, 13 and 14. The numerical data obtained from these curves are shown in Table III. The dotted portion of the curve was obtained by experiment and not by inference because in each case the complete load-deformation curve was obtained. The tensile strength and per cent total elongation are also included in Table III. It is to be noted that since the steels are of almost exactly the same thickness and the tensile specimens of the same

dimensions, the stresses are nearly proportional to the loads obtained for all of the steels. The load-deformation curves for any one of the steels may, therefore, be compared directly with those of any other. The curves were drawn on coordinate paper so that the different values could be read directly. They were then traced from the originals and reduced in size by photography.

Yield Point Elongation

Our experiments show that the yield point elongation increases as (1) the grain size of the steel is made smaller and (2) as the rate of deformation increases. This is shown in Figs. 15 and 16.

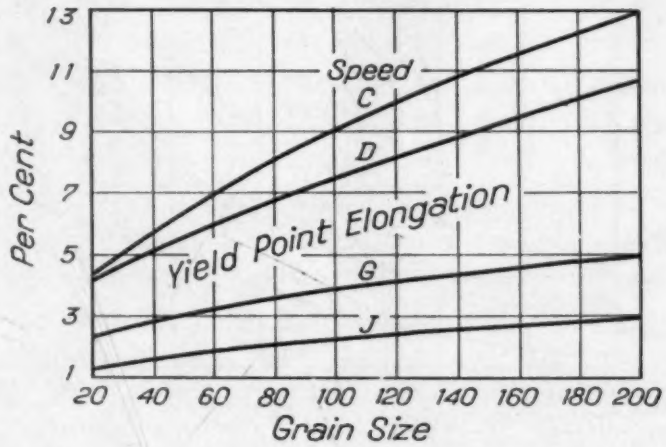

Fig. 15—Yield Point Elongation Versus Grain Size for Different Constant Rates of Deformation. (Grain Size in Number of Grains Per Square Inch at a Magnification of 100 Diameters.)

The phenomenon of uneven flow occurring during the yield point elongation can easily be followed with the eye during the elongation of the test piece. The formation of the first stretcher strain, however, occurs so rapidly that the eye can scarcely follow it. As the stretcher strains are formed and as the deformation continues, the depressions lengthen and broaden. The portions of the specimen after having undergone slip do not slip further until all other parts of the specimen have been so affected. From this it is clear that at the moment slip occurs, the rate of strain in inches per inch per minute becomes very great because the gage length in which plastic deformation is occurring is exceedingly small. It follows also that the extent of the yield point elongation is also dependent to some extent upon the number of stretcher strains existing at a

given instant. That is, the greater the number of stretcher strains present at the same time, the shorter will be the ultimate yield point elongation. Likewise, in the forming of the steel in the deep drawing operation, the actual rate of strain is dependent in a large measure upon the number of stretcher strains existing at a given instant and this, in turn, is, for the most part, dependent upon the presence of concentrations of stress resulting from the usually irregular contour

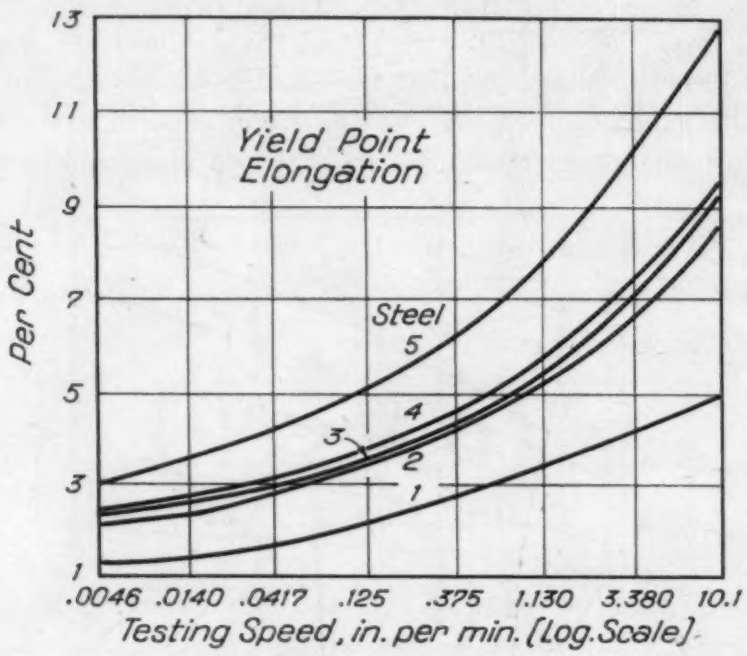

Fig. 16—Yield Point Elongation Versus Rate of Deformation for Steels of Different Grain Size.

of the stamping. It is clear, however, that the lower the rate at which a given deformation is produced and the larger the grain size of the steel, the less is the tendency for stretcher strains to occur and the less is the total deformation necessary to cause them to disappear.

As a consequence of the increase in yield point elongation with decreasing grain size, too small a grain size is, obviously, undesirable in steel for deep drawing purposes because of the greater tendency for stretcher strains to occur. In addition, since stretcher strains can in the tensile test only be prevented from occurring by elongating the steel through the yield point elongation, this degree of overstrain should be proportional to the amount of cold work by cold rolling or roller leveling necessary to prevent the occurrence of stretcher strains. The conclusion may, therefore, be drawn that the smaller

the grain size, the greater the amount of cold work necessary to prevent stretcher strains from occurring. This leads to another reason for a large grain size rather than a small grain size in steel for deep drawing: a given amount of reduction by cold rolling lowers the ductility of a fine-grained steel to a greater extent than the same amount of reduction lowers the ductility of a larger grained steel. This consideration should not be confused with the seemingly contradictory observation that "a pass on the cold rolls will not lower the ductility of a small grained sheet so much as it will a large grained sheet." These statements are not inconsistent because the *actual reduction* obtained per pass on the cold rolls is greater for the larger grained steel than the smaller grained steel. Or, to state this differently, less pressure on the cold rolls is required to reduce the larger grained steel a given amount than is required to reduce a small grained steel the same amount.

In this connection, it should be pointed out that the cold working by roller leveling or cold rolling does not necessarily have to be of such an amount as to cause the entire removal of stretcher strains in the tensile test before the steel is in a suitable condition for the deep drawing operation. This is because the deep drawing operation itself may be of a sufficient amount to pull the steel through the decreased yield point elongation. It follows from this, then, that the exact amount of previous cold work necessary to prevent stretcher strains from occurring in the finished stamping may be different for each individual shape or contour.

It is as undesirable, however, for the grain size of steel for deep drawing to be too large as it is for it to be too small. This is not only because the total elongation is lower as may be seen in Table III, but also because the coarseness ("orange peel") of the surfaces of the metal after drawing is directly proportional to the grain size. Experience has shown in this connection that a grain size of some 84 grains per square inch at a magnification of 100 diameters (corresponding to a grain diameter of 1/8 of an inch at the same magnification) is neither too large to "draw coarse" nor too small to present too great a difficulty in preventing stretcher strains from occurring by roller leveling on machines especially designed to bend the metal drastically.[11] This grain size corresponds to a grain size of 7 on the scale of the American Society for Testing Materials. It

[11]See U. S. Patents Nos. 1,649,704, 1,649,705 and 1,649,706.

should be pointed out, however, that a uniform grain size is also important even though the average grain size may be of the order just mentioned. This is because experience has shown that whenever a mixed grain size is present the small grains appear to predominate in their effect. There are, perhaps, a few instances such as fenders for automobile bodies when even the slightest change in the surface smoothness of the finished stamping is considered to be objectionable, and in these cases a slightly smaller grain size appears to be indicated. For most purposes, however, the grain size just mentioned has proven to be the most desirable.

The susceptibility of low carbon steel sheet or strip to "kink" ("break" or "flute") is greatly influenced by the grain size of the steel and by the rate of deformation. "Kinks" are merely localized stretcher strains and sometimes occur in handling ("coil breaks") or as a result of some bending operation such as roller leveling or in the manufacture of tin-plate cans. Their extent and number are very often of such a degree that even a subsequent deep drawing operation will not effect their removal. Furthermore, if any aging is permitted to take place in the steel after kinking and before deep drawing, the steel may even be deformed until fracture occurs without removing the effects of the previous local deformation. In this case, markings appear as vein-like elevations on the surfaces of the metal. The persistent greater resistance to plastic flow shown by these regions has, of course, been induced by the greater amount of aging which occurred as a result of the cold work attending the formation of the original local deformation. The peculiar susceptibility of low carbon steel to deform locally ("kink") may easily be demonstrated by bending small strips of the metal with the hands. The influence of a rapid rate of bending and a small grain size in intensifying this effect is most striking.

Yield Point

Our experiments show, also, that when specimens of annealed low carbon steel are deformed at different constant rates, the resistance to permanent deformation increases as the rate of strain is increased (see Fig. 17). They also show that the smaller the grain size, the higher is the yield point (see Fig. 18). In selecting the value for the yield point from the various curves, we have followed the views of most experimenters both here and abroad which are that

the "lower" yield point is the significant value rather than the "upper" yield point. The latter is believed to be due to certain characteristics of the tensile specimen. The value for the "lower" yield point is obtained from the load-deformation curve as follows: A line extending to the end of the yield point elongation is drawn perpendicular to the load ordinate at such a level that the areas formed above and

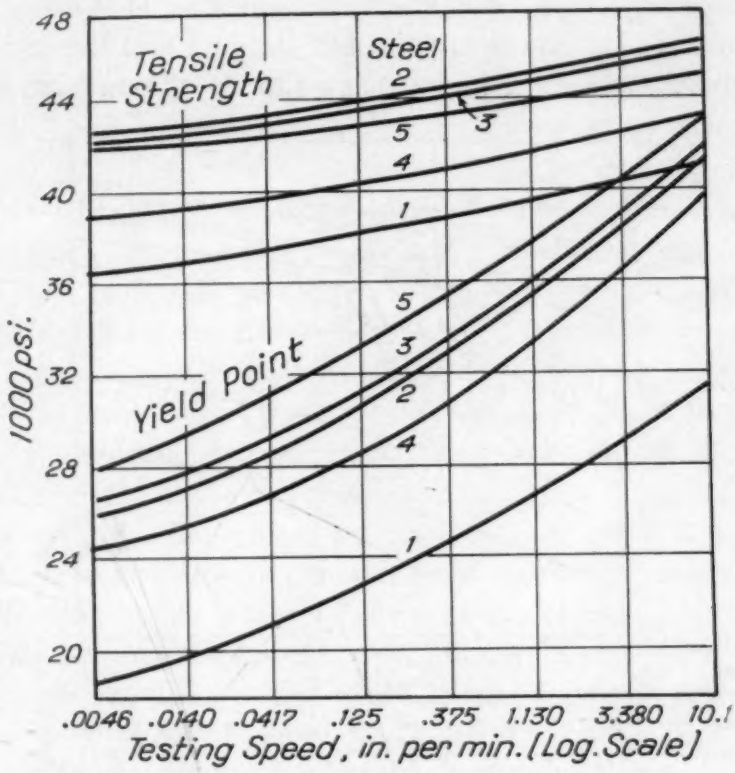

Fig. 17—Yield Point and Tensile Strength Versus Rate of Deformation for Steels of Different Grain Size.

below this line by its intersections with the load-deformation curve are equal. The intersection of this line with the load ordinate is chosen as the yield point.

Both of these observations are in accordance with generally accepted views, but the magnitude of the increase, particularly in respect to the effect of increasing the rate of strain, is worthy of particular notice. At the higher rates of strain, e.g., 1.5 and 4.44 inches per inch per minute, the yield point is within only a few thousand pounds of the tensile strength. It is significant that the value for tensile strength is, in comparison, only slightly affected (see Fig. 17).

A low ratio of yield point to tensile strength (called also the

"elastic ratio") is important in deep drawing because the sections of maximum stress in the stamping will develop only a certain maximum load before fracture will occur. The load at these parts must be sufficient to cause plastic deformation to occur in the succeeding larger sections because otherwise plastic deformation would not continue. On account of this, the lower the elastic ratio, the larger is the succeeding section which can be plastically deformed. In this connection, it is important to point out the fact that the ratio of yield point to tensile strength becomes greater as the rate at which the

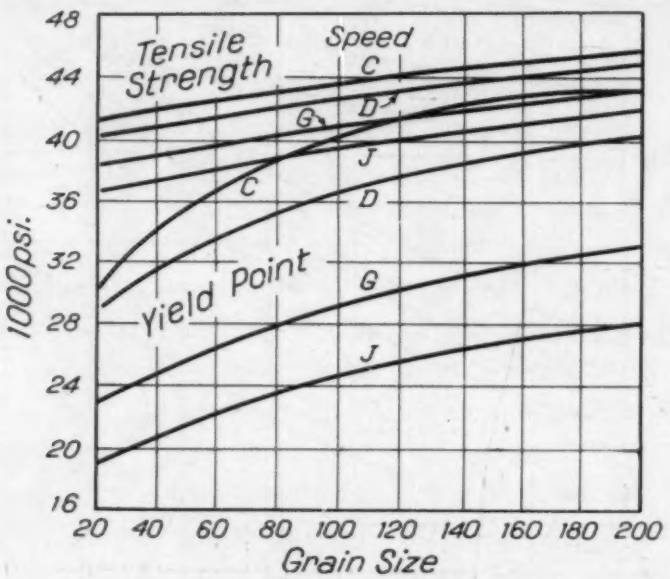

Fig. 18—Yield Point and Tensile Strength Versus Grain Size for Different Constant Rates of Deformation. (Grain Size in Number of Grains Per Square Inch at a Magnification of 100 Diameters.)

deformation is carried out increases. And, further, the steels of a smaller grain size are more affected by an increase in this rate than the steels having a larger grain size. The great influence of the rate of deformation on the ratio of yield point to tensile strength serves to answer in some measure the query existing in the minds of many as to the reason why this ratio has not appeared to be a better indication of the deep drawing properties of low carbon sheet and strip steel.

A high yield point is not desirable in steel for deep drawing, also, because of the greater inherent "toughness" of the steel. An approximate measure of the toughness may be expressed by the area under the stress-strain curve and the higher the yield point the larger this area becomes. Experience has shown that the greater the

toughness, the greater the tendency for buckles to occur during the deep drawing operation and, because of the greater loads necessary for deep drawing, greater wear on the dies will result.

As a peculiar consequence of the high yield point which can be obtained by rapid deformation, the condition might arise in which the yield point would become greater than the tensile strength. In this case, a reduction in thickness would take place in the part of the specimen first to slip, i.e., the first stretcher strain, which would be equal to the thickness of the sheet. In effect, sufficient work-strengthening to withstand the imposed load would be impossible and the specimen would fracture before other parts of the parallel section of the test piece had reached their respective yield points. This, of course, does not mean that the stress at the yield point can become greater than the stress necessary to cause fracture, but simply that the load shown by the specimen at the yield point is greater than the total load which can be carried by the specimen. Premature fractures of this type have actually been produced by one of the present authors.[12] The rates of strain in the ordinary deep drawing of sheet and strip steel are not, in the authors' experience, however, sufficiently high to cause this type of fracture. In passing, however, it is thought that the delayed transition of the steel from the elastic to the plastic state due to an increase in the rate of strain might possibly influence the results of impact tests or the behavior of the metal when called upon to withstand high loads suddenly imposed for short periods of time.

Method of Manufacture

Provided the grain size is the same, the various methods for rolling steel, insofar as these experiments are concerned, do not appear to affect differently the deep drawing properties of low carbon steel.

Discussion

There have been several theories advanced to explain the peculiar behavior of iron and low carbon steel as it first undergoes plastic deformation, but none of these, in the authors' opinion, have

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[12]Winlock and Lavergne, "Some Effects of the Amount and Rate of Deformation on a Low Carbon Strip Steel." Transactions, American Society for Steel Treating, Vol. XXI, Feb., 1933, p. 109.

been substantiated by any convincing experimental data. Most noteworthy, perhaps, is the theory held by some (principally in Germany) that the carbide of iron (Fe_3C) forms a network around the grains which breaks suddenly under the imposed load. The most which can be said for this theory is, that, in some respects, the behavior of the metal appears as though such a condition existed. It seems to us untenable because the purest irons obtainable which contain only negligible amounts of carbon show stretcher strains during plastic deformation.[13] That the phenomenon is associated with the grain boundaries is not necessarily true even though stretcher strains usually do not occur when the size of the grains is very large.[14] It would be necessary, first, to show that they would not occur when the rates of deformation are high.

The view has long been held that there are two opposing forces taking place when the resistance to permanent deformation is first overcome, i.e., at the yield point. The first is a reduction in area and the second is an increase in strength due to the cold working. It can readily be seen that the first is inherently a weakening process whereas the second is a strengthening process. From this point of view, it is clear that if the rate at which the first of these two forces takes place is predominant in its total effect, localized elongation will occur, but if the rate at which the second takes place is greater, general elongation will occur. Since work-strengthening can occur only after plastic deformation has taken place, the rate of work-strengthening is defined as: the rate of increase in unit stress per unit deformation. The conclusion is then drawn that the rate of work-strengthening is low during the yield point elongation because of the low slope of the curve drawn between the stress existing immediately before plastic slip occurs and the stress existing immediately after the first stretcher strain has formed. This latter stress may be readily obtained from the fact that the total yield point elongation is proportional to the reduction in area which has taken place.

This point of view does not appeal to the authors. It seems that it is just as fallacious to use as the basing point for the work-strengthening curve the stress existing at a point which can be in-

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[13]Private communication from R. L. Kenyon. See also, discussion by R. F. Mehl: "Aging of Metals and Alloys" by A. Sauveur. TRANSACTIONS, American Society for Metals, Vol. XXII, No. 2, February, 1934, p. 114.

[14]R. Arrowsmith, "A Note on the Effect of Grain Size on the Extension at the Yield Point." *Journal*, British Iron and Steel Institute, 1924, No. 2, p. 317.

creased or decreased simply by changing the rate of strain as it would be to take the temperature to which a metal can be undercooled as its freezing point. In our opinion, it would be more correct to use as a basis the stress existing at the point where it would occur if the rate of strain were extremely low. If this stress were accepted as being the yield point, and the true stress curve drawn connecting the stresses existing at the ends of the different yield

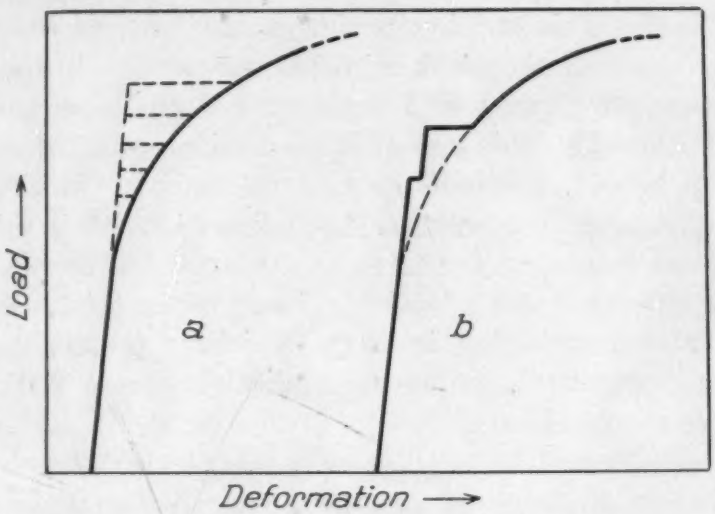

Fig. 19—(a) True (?) Load-Deformation Curve of Low Carbon Steel Showing, Also, Yield Points and Yield Point Elongations Which Are Obtained When Specimens Are Deformed at Different Constant Rates of Deformation. (b) Rate of Deformation Changed to a Higher Rate When the Yield Point Elongation Has Been Only Partially Completed.

point elongations obtained by increasing the rate of deformation, it would seem to us that the slope of this curve would more nearly approach the work-strengthening rate of the metal. To state this differently, if the cause of the increased resistance to permanent deformation produced by increasing the rate of strain were removed from the steel, the load-deformation curve would conform to the solid line in (a) of Fig. 19. That this hypothetical curve is of some significance is shown by the fact that it is the load-deformation curve of a steel which has been very slightly cold-worked as, for example, after roller leveling. In other words, a very slight amount of cold work does not appear to increase the ratio of work-strengthening to weakening (reduction in area), but simply causes a removal of the yield point. In our opinion, the elongation at the yield point occurs as a result of the unexplained phenomenon which causes the increased resistance to permanent deformation, rather than as a

result of an unexplained phenomenon which causes a "low" work-strengthening rate.

There is another peculiar phenomenon of interest in connection with the effect of the rate of deformation on the yield point elongation. If the specimen is elongated at some given moderate rate of deformation to a point where plastic deformation has only partially occurred and the rate of deformation then changed to a higher or lower rate, the yield point elongation continues to occur at a higher or lower load. This striking performance seems to indicate, also, that if the cause of the yield point phenomenon were removed, the load-deformation curve would follow the dotted line as indicated in (b) of Fig. 19. This curve simply designates, as does also the solid curve in (a), extrapolations of the rates at which the yield point elongation becomes smaller. Of further interest in this connection, hardness readings were taken at the bottom of stretcher strains formed at different rates of deformation. When these were plotted against per cent reduction in area (which is proportional to the yield point elongation), the curve obtained indicated that the metal hardened at a normal rate.

The reasons for the peculiar exaggerated "lag" or "hysteresis" in the initiation of plastic slip, is, then, of particular significance and importance in any study of stretcher strains. Their occurrence is, we believe, due to this delayed transition from the elastic to the plastic state. This view may be enlarged upon somewhat as follows: In any crystalline aggregate such as steel, there are certain to be some crystallographic plane or planes which are so situated that when a tensile load is imposed upon the metal, slip will occur upon these planes before it occurs on others. Immediately after slip has taken place, a concentration of the stress occurs at the parts first to slip due to the notch effect and the reduction in area. This stress is greater in magnitude as the stress which existed before slip took place is increased. Concomitantly, the rate of strain in inches per inch per minute increases. From this it follows that the lower the stress at which slip first occurs, the less the subsequent intensity of the stress, and, consequently, the less the necking (stretcher strains). The inherent work-strengthening properties of low carbon steel do not appear to influence greatly the extent to which such necking can continue. From the point of view of the ratio of work-strengthening to "weakening" (increase in stress due to reduction of area), it appears to us that the rate or rates of work-

strengthening remain constant for any given steel and that this rate is substantially independent of the stress at which plastic slip commences. In other words, as the stress necessary to initiate plastic slip increases, the tendency for necking to take place increases also. Stretcher strains occur as a result of this and not because of a low work-strengthening rate. That stretcher strains do not occur in freshly cold-worked steel is for the same reason that they do not occur in nonferrous alloys such as copper and aluminum, namely, because, as is well known, plastic slip occurs at comparatively low loads. In this connection, differential slip occurs in these metals also, but the depths of the stretcher strains are so extremely shallow that their appearance is referred to merely as a "slight roughening of the surface."

Delayed transitions and transformations of one kind or another are a common occurrence in metals, as, for example, the undercooling of a molten metal (already mentioned) or the lowering in steels by rapid cooling of the temperature at which the allotropic change takes place. Such phenomena are explained on the ground that insufficient nuclei are present to afford the necessary starting points. In the case of low carbon steel, a slight amount of cold work appears to produce the necessary nuclei from which plastic slip may start.

DISCUSSION

Written Discussion: By G. N. Schramm, Manager, Sheet and Strip Division, Metallurgical Department, Pittsburgh District, Carnegie-Illinois Steel Corporation.

We have read the paper by Messrs. Winlock and Leiter with a great deal of satisfaction and comfort in the thought that there are others who fully realize and appreciate the difficulties confronting the steel manufacturer in his endeavor to supply the steel fabricator with material which will not only meet all of the physical requirements but satisfactorily withstand the deformation required to produce a fabricated part. We are also grateful to the authors for the description of their findings, because the results of their investigations should be especially helpful to those manufacturers who appreciate that a customer not only knows what he wants but should receive it.

Lüder's lines, commonly known as stretcher strains, have been exceedingly troublesome for many stamping manufacturers. It is emphasized that by carefully regulating the amount of cold work done immediately before deep drawing, the load-deformation curve can be made relatively smooth from origin to fracture, without perceptible loss of ductility. Obviously, the smoothness of the curve indicates a satisfactory ratio between load and deformation.

The authors have described the influence of grain size on the yield point

elongation, in which it is shown that a very small grain size is undesirable in steel for deep drawing purposes, because of the greater tendency for stretcher strains to occur. They conclude that the smaller the grain size the greater the amount of cold work necessary to prevent stretcher strains from occurring. Obviously, therefore, the grain size should tend to be large rather than small, although not large enough to cause coarseness or "orange peel" of the surface of the steel. They mention the observation that coarseness after drawing is directly proportional to the grain size, and it is concluded that the optimum size is approximately 84 grains per square inch at a magnification of 100 diameters. Grains of this size are said to be neither too large to "draw coarse" nor too small to present great difficulty in preventing stretcher strains from occurring. This grain size corresponds to a grain size of 7 on the scale of the American Society for Testing Materials. Steel having this grain size can be roller leveled on machines especially designed to bend the metal drastically to prevent stretcher strains.

In producing satisfactory deep drawing steel, which can be subsequently finished in such a manner as to be free from stretcher strains, steel manufacturers may well aim, therefore, for a uniform grain of approximately this size. It may be well to point out, however, that the experimental steels used in this investigation had all received the same final heat treatment, namely 1250 degrees Fahr. (680 degrees Cent.), in a sealed container, and cooled slowly in the furnace. It is well recognized that the removal of such variables as the effects of cold work and of the differences in the rates of mill cooling was necessary for the proper conduct of the investigation. However, it is also well known that different mills, because of their peculiar economical limitations, treat their deep drawing steels by different methods, and that slight differences in manufacture, although producing approximately the same grain size, may have a very great influence on the susceptibility to stretcher straining. At the present time, therefore, it does not seem logical to expect steel coming from different sources to have identical properties even though the grain size in the different lots is approximately the same.

The effect of the speed of testing on the yield and ultimate strength of steel is again pointed out, both yield and tensile increasing as the testing speed increases, and the yield point increasing relatively faster than the tensile. This tends to bring the yield and tensile closer; in other words, it tends to increase the elastic ratio which is undesirable for deep drawing purposes. The authors state that, as a peculiar consequence of the high yield point which can be obtained by rapid deformation, the condition might arise in which the yield point would become greater than the tensile strength, and that premature fractures of this type had actually been produced by Mr. Winlock. It may be interesting to mention that we have observed this peculiarity on an unusual lot of fine-grained steel where the yield point was higher than the ultimate strength.

While the uniform grain size chosen as most suitable for the deep drawing requirements of the type of work mentioned by the authors, it does not necessarily follow that this grain size is best for all manufacturers of different kinds of stampings. Certain fabricators require different grain sizes for different fabricated parts.

Written Discussion: By H. S. Rawdon, National Bureau of Standards, Washington, D. C.

In a study of the surface markings which form in sheet steel when stressed sufficiently, carried out by the writer several years ago, to which the authors have kindly referred in their footnote 6, some observations were made on the microstructural changes of the metal which have a bearing on some of the points in the authors' discussion. These structural changes cannot be revealed by ordinary microscopic technique. It is necessary that the surface of the specimen be given a metallographic polish and that the grain structure be revealed by etching before it is strained. In Fig. 1 is shown the microstructure of a thin (0.0025 inch) sheet of annealed low carbon steel (0.23

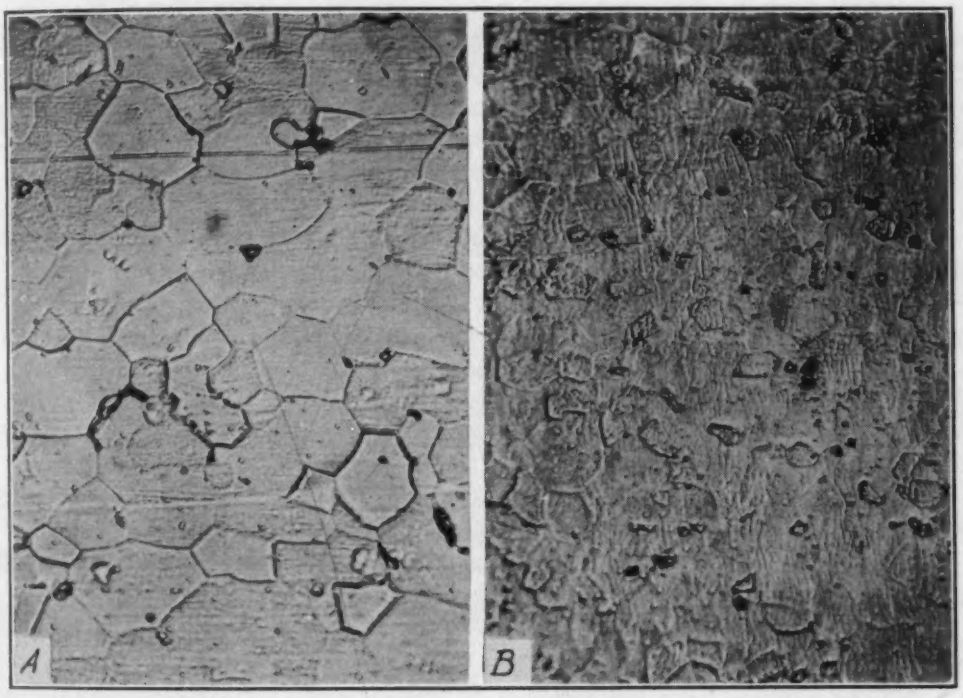

Fig. 1—Structure of Low Carbon Steel After Straining. × 250. A and B Represent Areas within a Strain Marking but not Identical Areas. A is Surface Polished and Etched After Straining. B is Surface Polished and Etched Prior to Straining.

per cent carbon) strained in tension sufficiently to produce strain markings. The area shown in both cases is within a strain-marking, specimen A having been prepared by ordinary metallographic technique, whereas A was polished and etched to reveal grain structure prior to straining. By this means, evidence as to the nature of the plastic flow of the material has been obtained. Nothing has been observed under such conditions which would support in any manner the hypothesis that the rupturing of an enveloping cementite film around the grains is involved in the flow of the metal resulting in the so-called Piobert or Lüders lines.

The ferrite grains are involved in the change but not all of them. Each grain which responds shows on its exposed surface a series of parallel mark-

ings which are the traces of the internal planes of slippage. These traces show a close parallelism in all the different grains affected and all are approximately perpendicular to the axis of stress application, at least, in specimens stressed in tension. Mehl and Gensamer[1] have shown by the back reflection X-ray method that the slipping occurred on the {3 1 0} planes of alpha iron.

It would appear then that only those grains which are oriented very similarly, with respect to ease of slip, can respond to the stress acting. On further consideration one accepts without question the postulate that there must be enough of these favorably oriented grains to form a fairly complete skeleton or network throughout the mass of the specimen under stress. A few isolated "favorable" grains cannot deform under any static stress, prior to the deformation of the entire mass, on account of the restraint offered by the surrounding "unfavorable" grains. The authors' correlation of the tendency toward stretcher-straining with fine grain size is in thorough agreement with this statement. In fine-grained material, the number of "favorable" grains is sufficient to form a continuous skeleton network which can "give" under the applied stress. In a steel having a relatively large ferrite grain size, the chance that there will be enough "favorable" grains to form such a continuous skeleton is very slight.

One phase of the phenomenon of strain markings which is often overlooked is the constraining action of the "unfavorable" grains. A moment's consideration, however, suffices to show that slippage within the "favorable" grains cannot occur unnoticed by the "unfavorable" neighboring ones. Some movement of the latter grains must occur if elongation from slip is to take place within the "favorable" grains. Such a movement, however, affects each grain or group of grains as a whole and is usually of the nature of tilting. Incidentally, it is to this that the characteristic matte appearance or dull finish of the surface of the strain markings is due.

These two components of the phenomenon are closely related to the "two opposing forces" referred to by the authors, the slipping within the "favorable" grains being the one force tending toward reduction in area and the opposition to deformation by the restraint offered by the "unfavorable" grains, the other. To one who examines the behavior of steel under stress from the structural standpoint, this explanation would seem to be the most satisfying.

As the authors have pointed out, it is now common practice to prevent the formation of stretcher strains by preliminary slight cold working of the sheet. On first thought, one is tempted to conclude that this is simply what is to be expected and that an opportunity is afforded in this way for the tendency of the "favorable" grains to early slipping to be accommodated. In addition, a general strengthening of all the grains would be expected to result. The explanation, however, is not so simple. This remedy is of a somewhat transitory nature and the original tendency of the sheet toward stretcher straining again manifests itself as the material "rests."

This is true for most low carbon steels although, of course, there is good reason for believing that there are some steels for which it is not true. The recovery of the tendency toward stretcher straining is intimately tied in with the aging of the steel. It is evident, as the authors have pointed out, that the freshly worked grains are so nearly alike in properties that differential strain-

[1] *Metals and Alloys*, Vol. 6, p. 158, June 1935.

ing cannot occur. Strengthening by aging, however, greatly improves the directional properties of the grains, at least many of them, and the preferential tendency of certain favorably oriented grains toward early slip again shows itself. There is no good reason for thinking that these are the same grains that showed this tendency before cold working. On the other hand, it seems reasonable to assume that they are not. The tendency of the so-called "favorable" grains toward early slip was undoubtedly fulfilled during the cold working but other grains or fragments were displaced from their original orientation and brought into such positions as to make them the "favorable" grains.

The writer wishes to congratulate the authors on the excellent results which they have presented. The facts established as to the effect of speed of deformation on the magnitude of the yield stress and the extraordinary difference in the yield point elongation produced by varying the rate of stress application will play an important role when the full explanation of the cold plastic deformation of metals is written. Similar data on steels stabilized by slight cold working and on the same after they have recovered their original tendency toward stretcher straining by aging would be a valuable contribution.

Written Discussion: By Reid L. Kenyon, American Rolling Mill Co., Middletown, O.

The authors are to be congratulated on their excellent presentation of this further contribution to our knowledge of the "yield point" behavior of ferrous materials. In studying the effect of speed of deformation, they strike at the very fundamentals of the problem and the data presented fill an important gap in the known facts concerning this interesting phenomenon.

The special testing machine used in this investigation was ingeniously conceived and cleverly constructed. It provides both very slow and very uniform speeds of deformation—a combination which is not always easily attained with many tensile testing machines in common use. There is a question which arises in this connection, and that is, the variation in load with the "head" motion of the machine. When an annealed mild steel specimen is stressed to the yield strength, there is a sudden yielding which permits a certain small movement of the head of the machine. This simultaneously releases part of the stress on the specimen, depending on the relationship between the movement of the head or the weighing mechanism and the load. In other words, the amount of "yield point elongation" depends somewhat on this relationship.

The data in Table III indicate a maximum in the per cent total elongation for rates of strain of about 0.018 to 0.056 inch per inch per minute. This is rather unusual in comparison with previous work that has been done on effect of rate of deformation on tensile properties. It would be of interest to know if the authors attach any importance to the fact that this occurs in each of the five steels tested, or whether they consider it to be merely a coincidence.

The authors have pointed out that even when the testing machine is operated at a constant speed to give a certain rate of strain, there are times during the test when deformation is taking place at much greater speeds. This is the case during the "yield point elongation" when deformation is highly localized in one or more stretcher strains. They further develop the idea that since the amount of elongation necessary to pull through the yield point depends on rate of deformation, the number of stretcher strains which form and spread simul-

taneously must greatly influence the behavior of the material both in the testing machine and in the drawing operation. This appears to be a logical deduction and one that might easily be subjected to verification by experiment. If one stretcher strain formed in the fillets at one end of the gage length of a tensile specimen and spread steadily down the specimen to the other end during the yield point elongation, the rate of deformation would appear to be only about half as much as in a case where stretcher strains started from both ends and met in the middle of the gage length. A two-to-one change in rate of deformation should, according to the data in this paper, result in a measurable difference in yield point elongation. The appearance of the stress strain curves in Figs. 10 to 14 indicate that both of these types of stretcher strain formation occurred on the various samples. I should like to ask if the authors have checked the above deduction against the variations in stretcher strain formation.

While it is a relatively unimportant point in relation to the main purposes of the paper, attention is called to the slightly different definition used by the authors for the "lower yield point." A number of years ago Bach proposed this term for the "lowest value to which the stress falls during the yield point elongation." This seems more rational than the "average" value used by the authors, although this is admitted to be somewhat a matter of opinion.

In discussing the mechanism of stretcher strain formation, the authors have taken exception to the theory of the breakdown of a network of a brittle constituent. While I believe they are fully justified in rejecting this theory as an explanation of this phenomenon, I am also unable to accept their explanation as given in the closing pages of the paper.

In a question of this sort, it is important to remember first of all that the determination of any of the "mechanical" properties of materials involves not only the behavior of the material itself, but the machine used in performing the test. The result obtained is influenced by the material, the size and shape of the specimen, the testing machine and the method of carrying out the test. The results of such tests are therefore of value only when they can be correlated with actual service conditions. It has been shown by G. Welter in the March 1936 METAL PROGRESS, for example, that the yield point behavior of mild steel can be greatly affected or entirely eliminated by modifying the conditions of the test.

The behavior of mild steel at infinitely slow rates of loading may be of some theoretical interest, but certainly does not represent the conditions of service under the drawing press. While it might be possible to deform mild steel slowly enough to avoid stretcher straining, this would be too slow for drawing shop operations.

Even when considered from a purely theoretical angle, the behavior at zero rate of strain does not give a satisfying explanation of the yield point elongation which occurs in mild steel under practical conditions of testing and use. Although the careful and interesting work presented by the authors contributes materially to our knowledge of yield point behavior, much further work remains to be done to explain the mechanism which causes this phenomenon. The stages in deformation postulated on page 184 of the paper would occur in nonferrous as well as ferrous metals, and take no account of

the disappearance of the stretcher strain behavior in mild steel when the test is conducted at slightly elevated temperature or when the grains are unusually large. It has also been observed that very fine-grained steels may also be free from yield point elongation so that there appears to be some critical grain size which will produce maximum yield point elongation.

The question still to be answered is, why do we get such a region of instability in mild steel under certain conditions and not others? The final solution rests upon the accumulation of just such information as reported by the authors, and it is hoped this paper will prove a stimulus not only to them, but to others, to pursue a further study of this interesting subject.

Written Discussion: By Dr. Charles W. MacGregor, Instructor of Mechanical Engineering, Massachusetts Institute of Technology, Cambridge, Mass.

A number of curves are shown in Figs. 10 to 14 indicating that both the upper and the lower yield points have increased very markedly with increasing strain velocities. It can be seen from Table III that with a maximum speed ratio of about 2200 to one, the average per cent increase of the yield point for steels Nos. 1 to 5 is of the order of 57 per cent. The same table gives an average per cent increase in the tensile strength for all the steels tested, with the same maximum speed ratio of 2200 to one, of about 10 per cent.

Two or three years ago, H. Deutler[2] and H. Brinkmann[3] made very careful tests to determine the effect of testing velocity on the yield point and tensile strength of such ductile materials as mild steel and copper. In these tests the strain velocity was varied as much as in the ratio of a million to one. The mild steel used by Deutler was of the following composition:

C Per Cent	Si Per Cent	Mn Per Cent
0.20	0.30	0.60

It was found from Deutler's tests that with a speed ratio of 10,000 to one, the yield point increased about 24 per cent for this material, while the tensile strength increased only about 8 per cent. A comparison of these results with those shown in Table III of this paper indicates that the effect of speed of testing on the tensile strength as found by the authors is of about the same order of magnitude as that found by Deutler. The effect of testing speed on the yield point, however, is over twice as great as the effect obtained from Deutler's experiments. It is to be noted that the mild steels used in the present paper are of a much lower carbon content than the material used by Deutler which may account for the difference in the two results. It is well known that such ductile materials as lead and tin are much more affected by speed of testing than mild steel or copper. The results of the authors are therefore interesting in that they seem to indicate the possibility of a considerable speed effect variation even among mild steels of different carbon content. The writer feels therefore that the possibility of such an effect provides a fertile field for further investigation.

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[2]H. Deutler, "Experimentelle Untersuchungen über die Abhangigkeit der Zugspannungen von der Verformungsgeschwindigkeit," *Physikalische Zeitschrift*, 33, p. 247-259, 1932.

[3]H. Brinkmann, "Zerreissversuche mit höhen Geschwindigkeiten," Dissertation, Technische Hochschule, Hanover, 1933.

In some of the early tests on the effect of velocity of testing on material properties, P. Ludwik[4] suggested the law

$$S = S_1 + S_2 \log \frac{v}{v_1} \qquad (1)$$

where S = stress corresponding to the velocity of testing v, S_1, S_2, v_1 = material constants. This law is not valid for very low rates of flow since the equation obviously fails for v = O. However, equation (1) has been checked for larger rates of flow by Deutler, Brinkmann, Jameison, and others. The writer would therefore be interested to know whether or not the authors have found that this equation agrees with their results.

It is also mentioned in the paper that there is considerable doubt that the observed surface markings in copper, brass, and aluminum are of the same character as the stretcher strains which occur in mild steel. The writer has also observed such flow layers in aluminum and copper, both materials being in the work-hardened state.[5] It was found that by pre-cold working the aluminum, for example, rather pronounced flow layers could be obtained on subsequent stressing. The effect of the pre-stressing was to give the aluminum a more pronounced transition from the elastic to the plastic state. That these flow layers have a different appearance from the stretcher strains obtained in mild steel is quite true. A possible reason for the difference in appearance in the two cases may be seen when the stress-strain curves for the two materials are considered. As soon as a stretcher strain forms in mild steel, there is an immediate drop in load on the testing machine and the surrounding layers of material are not immediately over-stressed. In the case of aluminum, however, there is no drop in load and adjacent layers are also deformed giving a more uniform appearance. The mechanism of flow in both cases is probably the same, however, being due to slip on the planes of maximum shearing stress.

In applying analytical treatment to problems involving materials having stress-strain curves such as shown in Figs. 10 to 14, it is obvious that the plastician is amply justified in using for beginning yielding an ideal stress-strain curve such as given by the initial elastic line to the yield point followed by a horizontal line parallel to the strain axis. In many plasticity problems this idealization proves to be of considerable help in the mechanics of the solution.

Written Discussion: By Alvan L. Davis, Scovill Manufacturing Co., Waterbury, Conn.

This paper is of the outstanding excellence we have learned to expect from Mr. Winlock and his associates. It illuminates a subject which has been none too clear, both by the novel and arresting quality of the data given and by its lucid presentation. The close correlation of the test data with manufacturing practice gives especial value.

It is to check the observation as to acceptable grain size for cold forming operations, and possibly extend it somewhat, that this comment is presented;

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[4]P. Ludwik, "Uber den Einfluss der Deformations geschwindigkeit bei bleibenden Deformationen mit besonderer Berucksichtigung der Nachwirkungserscheinungen," *Physikalische Zeitschrift*, 10, p. 411-417, 1909.

[5]C. W. MacGregor, "The Formation of Localized Slip Layers in Metals," *Metals and Alloys*, Feb. 1933.

also, to make a plea for the use of the same terminology that has for years been employed for grain size in nonferrous metals.

As to optimum grain size for cold forming, pressing, deep drawing, etc., I check with the statement that 84 grains per square inch (at a magnification of 100 diameters) is excellent. The A.S.T.M. classification of grain size in steel is given in arbitrary numbers from 1 to 8, according to the number of grains per square inch. The following table gives the corresponding grain size as used for all other metals than steel, where the diameter of the average grain is given in millimeters.

A.S.T.M. No.	No. Grains Per Sq. In. at 100 ×	Corresponding Ave. Grain Size in MM.	Notes on Grain Size and Surface Rating After Cold Forming — On Basis of Smoothness and Freedom from Orange Peel
1	Up to 1½	0.207 mm	
2	1½ up to 3	0.147 mm	
3	3 up to 6	0.104 mm	
4	6 up to 12	0.073 mm	0.064 mm—very coarse and bad
5	12 up to 24	0.052 mm	0.052 mm—bad—rough surface
6	24 up to 48	0.037 mm	0.040 mm—poor 0.037 mm—mediocre 0.034 mm—fair to good
7	48 up to 96	0.026 mm	0.030 mm—good 0.027 mm—very good 0.024 mm—excellent
8	96 up to 192	0.018 mm	0.018 mm—exceedingly fine — O.K. for electro plate without buffing
Extension			
9	192 up to 384	0.013 mm	No experience
10	384 up to 768	0.009 mm	No experience

The five steels in the paper under discussion as classified are:

Steel No.	No. Grains Per 1 Inch Square	Actual Ave. Grain Size MM.	Report As to Suitability for Cold Forming with Satisfactory Surface
1	20	0.057 mm	Too coarse grained
2	84	0.028 mm	O.K.
4	91	0.027 mm	O.K.
3	107	0.204½ mm	O.K.
5	200	0.018 mm	Too fine—for ready suppression of stretcher strains

My comment is that the 0.028 mm to 0.024 mm grain size is exactly the range that we find gives best results for forming and deep drawing. For shallower stampings, the 0.018 mm may be sufficiently cold-rolled to suppress stretcher strains and still stand the forming operation. They will then have an unusually smooth surface, which will not require buffing before chromium plating. It is believed that the average diameter of grain as expressed in mm is simpler than the other system and lends itself to clarity of thought upon the subject — making the comparison easy with other metals.

Written Discussion: By V. N. Krivobok, Allegheny Steel Co., Brackenridge, Pa.

It is a pleasure to read a paper such as was given by Winlock and Leiter. The task of discussing it, equally pleasant, is by no means easy. The work has been conducted with such care, foresight and intelligent approach that the discussion resolves, of necessity, into praise and, perhaps, suggestions.

The intricacy of the studied phenomena can be readily appreciated when we consider the number of variables that influence the behavior of the cold-

rolled steels. Among such variables should be placed the rate at which a given amount of cold reduction was given to steel. The authors have considered the amount of cold reduction; this is obvious from their statement, which is quoted: "a given amount of reduction by cold rolling lowers the ductility of a fine-grained steel to a greater extent than the same amount of reduction lowers the ductility of a larger grained steel." It is not clear, although I carefully read the following paragraphs, whether the quoted observation will hold true regardless of the number of passes by means of which a certain amount of cold reduction is obtained, that is, the rate of deformation, which is not to be confused with the speed of deformation (the authors call the latter "rate"). To be more specific, it is realized that mechanical properties of stainless (austenitic type) alloys depend, to a marked degree, whether reduction from, for example, 0.080 to 0.015 inch was accomplished in 5 or 50 passes. While the phenomena occurring during cold rolling of metastable alloys, just mentioned, may be different, I feel that it is a pertinent suggestion, upon which the authors may wish to elaborate and relate their experience.

The experimental results show that the depth of the stretcher strains are proportional to the yield point elongation (page 167). Subsequent observations indicate that both the grain size and the rate (speed) of deformation have influence upon the extent of yield point elongation. The latter, judging by Figs. 15 and 16, is much more pronounced in its influence than the former. In fact, it appears quite pertinent to ask if the rate (speed) of deformation is not the deciding factor? Three steels (No. 2 with 84 grains per square inch, No. 3 with 107 and No. 4 with 91), when contrasted with Steel No. 1 (20 Grains) show the difference in yield point deformation from minimum of 1.35 per cent (steel No. 1) to maximum of 2.45 per cent (for steel No. 3).

For the finest grain steel (No. 5) we have yield point elongation minimum of 3.0 per cent. On the other hand, the values for this yield point elongation for the coarsest steel are from 1.35 to 4.4 per cent, dependent upon rate (speed) of deformation. No doubt Mr. Winlock and Mr. Leiter have considered these actual rather than relative values, and perhaps they would subscribe to the suggestion that the two factors which they studied are inter-related and that their finite influence is of more complete nature.

And while we are on this subject, I should like to be allowed to ask the authors for further interpretation of that part of their discussion which appears on page 175 of the paper. My interest is so profound and sincere that I do not wish to miss, unwillingly, any point of the discussion. To quote:

> "As the stretcher strains are formed and as the deformation continues, the depressions lengthen and broaden. The portions of the specimen after having undergone slip do not slip further until all other parts of the specimen have been so affected. From this it is clear that at the moment slip occurs, the rate of strain in inches per inch per minute becomes very great because the gage length in which plastic deformation is occurring is exceedingly small. It follows also that the extent of the yield point elongation is also dependent to some extent upon the number of stretcher strains existing at a given instant. That is, the greater the number of stretcher strains present at the same time, the shorter will be the ultimate yield point elongation."

The last phrase, intended I believe as a conclusion, is not quite clear to me. I

am afraid that I do not grasp the logic of the thought and, consequently, would be grateful for additional discussion.

Very interesting data are collected in Fig. 17 and also in Fig. 18. Since we are engaged in open and friendly discussion of valuable contribution, a following question, based on mentioned curves and the reasoning applied by authors to these curves, may simply be asked: does not the evidence presented indicate that the most important factor of deep drawing is the rate (speed) at which these operations are carried out? Several instances come to my memory when the adjustments for proper operating procedure have resulted in the markedly shown tendency to diminish difficulties and break away from the "stretcher strain" phenomenon.

Winlock and Leiter teach us, however, that grain size is also of some importance: the inadequacy of the word "some" is fully realized, but a better substitute is difficult to find.

Adjustments in the rate of deformation, we gratefully learn, may compensate for the influence of grain size, but the disclosed information brings again to mind, and rather forcibly, the desire to know the reasons for this complete interdependence between rate, grain size and available properties. We cannot be satisfied with mere statement, (fully upheld by experimental results) that "steels of a smaller grain size are more affected by an increase in this rate (i.e., rate of deformation) than steels having a larger grain size." The authors are obviously the men to satisfy our scientific curiosity, and I, for one, am looking forward with sincerest appreciation to their next contribution.

Written Discussion. By Albert Sauveur, Harvard University, Cambridge, Mass.

The subject of "stretcher strains" has received the attention of those engaged in the deep stamping of steel sheets for many years, and a fully satisfactory method of preventing their occurrence is still to be found. The authors of this paper have shown conclusively that a relatively large grain size and slow deformation are helpful, and their findings should be welcome and appreciated. Referring to their discussion of cold work deformation and work strengthening, I am not sure that I quite separate the two phenomena. They appear to me closely associated, the latter being the result of the former. Whenever cold work deformation occurs, work-strengthening must follow. Increasing the speed of deformation must, it seems to me, increase the speed of work strengthening; likewise increasing the magnitude of deformation must in a like measure increase the magnitude of work strengthening.

Oral Discussion

R. F. MEHL:[6] I am not going to meet Mr. Winlock's request for an answer to his questions, but I shall do the next best thing,—summarize the various facts that will have to be accounted for in furnishing an answer.

Several factors are known to control the occurrence of a yield point elongation. Iron and steel show it markedly, and are usually thought unique in this respect, but are not, for slight yield point elongations are shown by gold-silver and by copper-nickel alloys. It cannot be argued, as some have tried, that the yield point elongation in iron is brought about by the occurrence of a network

[6]Carnegie Institute of Technology, Pittsburgh.

or superstructure of precipitate or impurity and that this structure must be progressively destroyed at constant load before the true plasticity of iron can assert itself, for pure iron—or iron as pure as we can now make it—shows a yield point elongation just as marked as a mild steel, as work I performed several years ago clearly shows.

The yield point elongation may be eliminated in several ways: if the temperature of test is raised above 400 or 500 degrees Cent. (750-930 degrees Fahr.), the stress-strain curve no longer shows it; if steel is quenched from below A_1, it is no longer there; any method of treatment which will induce internal stresses, such as cold work, or induce stress concentrations, such as notches, will remove it. Furthermore, single crystals of iron show no yield point elongation, as Pfeil has shown.

Composition exerts an important effect also, for at carbon compositions greater than eutectoid no yield point elongation can be found, though unfortunately few stress-strain curves are available for a general study of the effect of composition—a study which would be particularly interesting for alloys in which the gamma field is restricted and the alpha extended.

As a matter of general interest, I think I should recount the visit I had several weeks ago from Professor Haigh, whose work in England on fatigue and the design of fatigue machines is well known. Professor Haigh stated that he had been able to obtain upper yield points in mild steel at loads as high as 118 per cent of the ultimate tensile strength, and that the yield point elongation and the rest of the tensile curves were in no way abnormal. This high value of the upper yield point was obtained by preparing a tensile test piece with the greatest of care, using a highly polished surface, fillets of very large radius, and by loading the piece in such a way as to approach true axiality. It would seem as though the load was increased so carefully that the steel had hardly become aware of it—much as a tax load may be increased on an unsuspecting body politic!

All these facts must be explained before we may feel fairly competent in dealing with such simple yet basic problems in the plasticity of iron and steel.

The present work by Messrs. Winlock and Leiter is a very acceptable contribution indeed—it will doubtless become textbook material, and a higher compliment hardly could be paid to the fundamental nature of any work. It is surprising that so little has been published by the producing industries on the plasticity of deep drawing steel sheets. There is a great deal still to be done, particularly with respect to the analysis of the mechanics of deep drawing. Even though the Watzau tester does not reproduce all of the deep drawing conditions, it seems to me that an extended study of the forming properties of steel sheet using this tester would be worthwhile indeed. Even though the promise of immediate practical results would not be great, industry should hardly expect phenomenal results from research and should be happy enough if long continued study should make industrial processes more readily understood and should effect minor improvements. Such a study could well be financed by the producing and consuming industries. Certainly this paper shows what knowledge high grade research can produce.

RICHARD F. MILLER:[7] Last year, at the Hammond Laboratory, Yale Uni-

[7] U. S. Steel Corp., Research Laboratory, Kearny, N. J.

versity, I carried out slow tensile tests on aluminum single crystals of two degrees of purity (99.95 and 99.90 per cent). Crystals were tested at a series of temperatures, from room temperature to 300 degrees Cent. A marked yield point extension was found to exist in both cast and recrystallized single crystals, beginning at the end of the elastic range and continuing with little increase of stress up to 1 or 1½ per cent permanent deformation. This yield point extension was more marked the purer the metal and the lower the temperature at which the test was carried out; it disappeared entirely above the lowest temperature of recrystallization (about 250 degrees Cent.). The phenomenon was also noted, though to a much less marked extent, in silver single crystals below the recrystallization temperature. It was not noted in zinc single crystals, probably because the tests were all made at room temperature or above, and zinc is known to recrystallize below room temperature.

Possibly the reason that the yield point extension has not been previously noted in nonferrous materials is because it occurs in the early stages of plastic deformation, and then only in pure materials and in a certain temperature range. However, the similarity to the behavior of annealed mild steel in tensile tests seems to be greater than can be accounted for by coincidence, and it may be that this behavior is a general characteristic of metallic crystals.

M. GENSAMER:[8] We have been studying the yield point in iron, and I think that a preliminary report of some of the things that we have found out should be interesting. In setting up our program, we decided to begin by studying single crystals of iron. We wanted to study the effect of orientation of the crystal on the yield point, if it really existed. It seemed to us that the only way to do this was to load the crystals extremely slowly, but we compromised by adding small increments of load and waiting a long time between increments.

The results of our experiments may be summarized as follows:

1. Single crystals of iron have a very definite and sharp yield point when studied by the method of applying dead weights in small increments, waiting several hours at each load before applying the next increment.

2. When slip does occur, using this method of loading, there is a well defined time lag between the application of the increment of load and the beginning of flow. It is necessary to wait several hours, with load increments of about 5 per cent of the yield point, before flow begins, using an optical strain gage with a sensitivity of better than 10^{-5} inches per inch. When flow begins it speeds up, then slows down and finally ceases, the specimen having reached a length which is constant, at least for a period of 5 days. On the application of the next load increment the specimen behaves the same way.

3. A polycrystalline iron specimen reacted the same way in this test. No flow was observed below a well defined yield point. On applying a small increment of load above this yield point, the iron flowed, but flow did not begin immediately on application of the load increment. The "induction period" or time lag was observed to be 14 hours in one case, somewhere over 24 hours in other cases, and in one case where a considerable increment of load was applied it was reduced to a few minutes.

This behavior of iron must be intimately related to the phenomena described

[8]Carnegie Institute of Technology, Pittsburgh.

by the authors, and I believe that it is, in fact, the underlying cause of them, although I do not yet see how it can explain the greater amount of yield point elongation at faster rates of straining. Although we may not be able to answer the questions raised by Mr. Winlock, we hope at least to simplify the problem.

Authors' Closure

We are greatly indebted to those who have discussed our paper for their extremely valuable contributions. It is only with the help of constructive criticisms such as these that real progress can be made. It is gratifying to us that our attempt to shed some further light on this important problem has met with their approval and we should like to thank them for their kind and generous remarks. If we have done no more than arouse the interest of metallurgists in one of the greatest difficulties which confronts the manufacturer and user of sheet and strip steel for deep drawing, we feel that a great deal has been accomplished toward obtaining a solution.

It would be quite impossible to reply to all of the questions asked or implied because as our paper clearly indicates, and as has been pointed out in the discussion, there is much work still to be done. No one realizes this more than we do. We agree entirely with Dr. Mehl that surprisingly little work has been done in the mechanics of deep drawing or in the metallurgy of the steels used in deep drawing. The reasons for this are not clear. For, as we pointed out in the opening paragraph of our paper, it is our opinion that "a constant or increasing rate of progress can be obtained only when equal attention and emphasis are placed on all phases of any development."

The pertinent comments which have been made necessitate a description of a few further experiments and it is hoped that the results of these, together with a further discussion of some of the points already mentioned in our paper will serve to answer in part the questions which have been raised and will perhaps explain more clearly a few of the phenomena which, as we see them, are occurring at the yield point of low carbon steel.

One of the phenomena which we believe to be of fundamental importance is the so-called "upper" yield point. A metal may be said to have an "upper" yield point when the stress to which it can be subjected without causing plastic deformation is greater than that stress to which it can be subjected immediately after the occurrence of plastic deformation without causing further plastic deformation. As will be explained, it does not necessarily follow from this that the load-deformation curve must show the presence of a load greater than the load at which the yield point elongation is taking place to indicate this higher load carrying capacity of the steel while it is in the elastic state.

Fig. 1a and Fig. 1b of this discussion show two load-deformation curves of the same steel tested at the same rate of deformation. They are somewhat different in appearance, but they are, we believe, the same insofar as the inherent properties of the steel are concerned. From O to A in curve (a) of Fig. 1, the deformation is substantially elastic. At the instant the load A is reached, the inherent load-carrying capacity of the metal is decreased and the load starts to drop. At the same instant that the load A is reached a stretcher strain is formed, whose depth is proportional to Aa. It is clear that since this first increment has been deformed at this load, the extent to which it has been

cold-worked is proportional to Aa. The first increment of metal to deform plastically cannot, therefore, deform further until the load a is again reached. As soon as plastic deformation has started in this fashion, the rate of strain is increased tremendously (even though the head speed of the testing machine remains constant) because the gage length of the metal now undergoing plastic deformation is extremely small, i.e., a small portion of the metal at each end of the stretcher strain. The increased resistance of the steel to con-

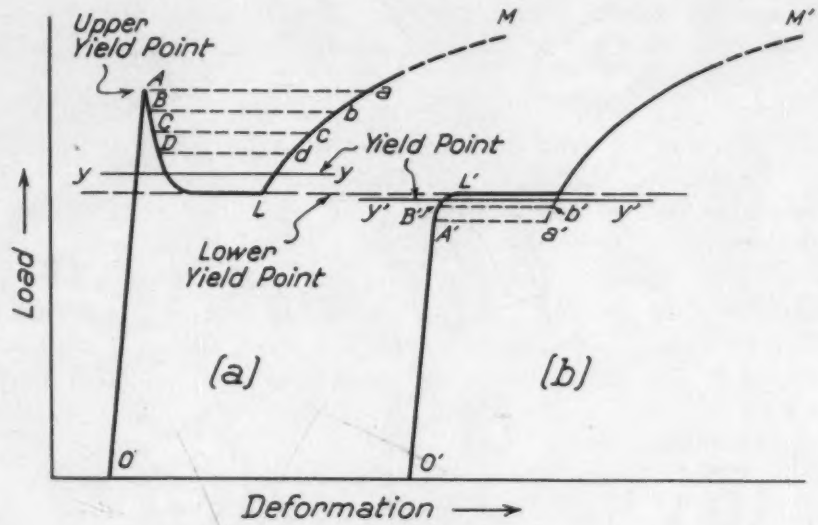

Fig. 1—Load-Deformation Curves.

tinued plastic deformation resulting from the increase in the rate of strain thus produced serves to slow up the rate of drop in the load. As plastic deformation proceeds and the load continues to drop, a reduction occurs at both ends of the first stretcher strain whose depth is slightly less, viz., proportional to Bb. This procedure continues at the lower loads C, D, etc., until the load L (the "lower" yield point) is reached, at which the greater part of the metal in the gage length is plastically deformed. The reason a constant load is reached is because the inherent tendency for the load to drop is exactly balanced by the effect on the metal of the increased rate of strain. That is, the amount which the load may drop during the yield point elongation and the extent of the yield point elongation is influenced by the rate of strain and by the work-strengthening characteristics of the steel. It should be emphasized, as we have shown in our paper, that the resistance to the initiation of plastic deformation and the resistance to continued plastic deformation at a given rate of strain are different for different conditions existing in the steel, e.g., grain size, temperature, etc.

When the end of the horizontal part of the load-deformation curve is reached, the metal continues to deform nonuniformly (but without further increase in the differences in level of the surface markings) under the loads designated by d, c and b. This continues until the whole gage length has elongated to a thickness equal to the depth of the first stretcher strain. At this moment, the load shown by the testing machine is at a on the load-

deformation curve. From a to M (the maximum load), the extension is uniform throughout the whole gage length. The curve L d c b a M is, of course, not constant for all rates of deformation, but was drawn in this manner to simplify the sketch. Actually, the points d, c, b and a lie on separate curves slightly to the left of the positions shown. We have not investigated, as yet, this portion of the curve as suggested by Ludwik's equation to which Dr. MacGregor refers. The load YY, as mentioned in our paper, is chosen as the yield point and is obtained by making equal the areas subtended by this line and the load-deformation curve. The value for yield point chosen in this manner is, then, the approximate average load at which the nonuniform elongation takes place.

An exceedingly important consideration, according to our view, is that the speed at which the different increments of metal travel along the dotted lines Aa, Bb, Cc, etc., in Fig. 1a and the length of these lines, i.e., the depth of the stretcher strains, are governed by the magnitude of the load existing at the instant plastic deformation is initialed. That is, the greater the load, the greater is the speed at which the deformation takes place, and the greater is the amount of deformation. In other words, as soon as plastic deformation is initialed, each increment of metal travels at a speed and for a distance which are governed by the load to which the metal finds itself subjected and by the resistance of the steel to continued plastic deformation, i.e., by its work-strengthening characteristics.

As Dr. Mehl so aptly points out, it would appear as if the metal were caught unawares by the high load upon it, and its subsequent behavior during the yield point elongation is controlled and governed by the magnitude of this inordinately high load and the resistance of the metal to continued plastic deformation. If the load at the instant plastic deformation starts is high, the speed of deformation is likewise high and, since, as has been described, the resistance to deformation increases with the speed of deformation, the load is not permitted to drop to the extent it would like.

In connection with this it is interesting to refer to the experiments of Professor Haigh mentioned by Dr. Mehl, in which it is stated that loads as high as 118 per cent of the ultimate tensile strength were obtained, and that the yield point elongation, and the remainder of the load-deformation curve were in no way abnormal. Our experiments indicate that if Professor Haigh had employed a rapid rate of deformation the condition of plastic flow would then be like that given on page 181 of our paper in which "a reduction in thickness would take place in the part of the specimen first to slip, i.e., the first stretcher strain, which would be equal in depth to the thickness of the sheet." In this case, fracture would occur almost immediately without the whole gage length having been plastically deformed. Under these circumstances, the shape of the load-deformation curve would not be normal, but would have the shape of a banner at the end of a flagpole in a light breeze. We have obtained such curves experimentally.

The type of curve shown in Fig. 1a can always be obtained in annealed low carbon steel of small or moderate grain size if (1) the edges of the specimen are carefully rounded by polishing; if (2) the fillets on the specimen

are large; if (3) great care is exercised in properly aligning the specimen in the grips of the testing machine; if (4) there are no indentations made by the extensometer; if (5) the surface of the specimen is free from pits, scratches or other surface imperfections; and (6) of greatest importance, if the specimen is free from any cold-worked material which, for example, might be present as a result of the machining operation, or because of some local abrasion, or by a bent specimen.

Such great care in the preparation of the specimen appears to be necessary and the determination of the exact location of the upper yield point appears to be so difficult that we offer to Dr. Gensamer the suggestion that the peculiar and interesting behavior of his single crystals of iron may be due to a slight raising of the stress due to jarring of the specimen during the test. The upper yield point appears to be as void of reason and as evasive as the body politic which Dr. Mehl has already excused as being unsuspecting! Some of our experiments indicate that the value of the upper yield point may be an inherent property of low carbon steel of a given grain size or heat treatment and is not influenced by speed of test. Of this, however, we are not sure. It would be interesting to know more of this and, also, to learn, as Dr. MacGregor points out, the effect of larger amounts of carbon on the yield point behavior of steel.

Referring briefly to Fig. 1b, if there is present in the specimen any metal which has been cold-worked or if there is anything which will cause a severe concentration of the imposed load, a load-deformation curve of this type will be obtained. The important observation should be made that the load at which the major portion of the yield point elongation occurs is, nevertheless, the same in both curves of Fig. 1. The curves which we obtained in our paper were almost all of this type and the cold-worked material and the concentration of stress necessary to cause the omission of an exhibited peak stress in the load-deformation curves were the result of the impressions made by the points of the extensometer. When load-deformation diagrams of this type are obtained, the first increment of the metal, according to our view, deforms plastically at the load A′ (Fig. 1b) and the depth of the first stretcher strain is proportional to A′ a′. We should like to emphasize the point that it is this horizontal portion of the curve which, as has been shown in our paper, is so greatly influenced by the rate of deformation,—not only in regard to the load at which it occurs, but also in regard to the extent to which nonuniform elongation may continue before uniform elongation takes place. The "yield point" is chosen from this curve in a manner similar to that shown in curve (a) of Fig. 1. It will be seen that the values for "yield point" are not quite the same, but it should be noted also that the values for percentage yield point elongation are not the same either, i.e., the percentage yield point elongation in curve (b) is slightly less than in curve (a). This method of choosing the value for "yield" point, to which reference has been made by Mr. Kenyon, appears at first glance to be questionable because it gives a slightly different value for the same steel tested at the same speed. The justification for choosing the value for yield point in this manner lies in the relation of this value to the percentage yield point elongation. As has been explained, the extent of the yield point elongation is influenced by the actual load on the

specimen at the instant plastic deformation starts. This load, in turn, varies largely in test specimens which have not been especially prepared. For example, in curve (a) of Fig. 1, uniform elongation does not begin until the load a is reached; but in curve (b) of the same figure, uniform elongation begins as soon as the curve departs from the horizontal and starts upon its upward course. It becomes necessary, then, in any attempted correlation of "yield point" and "yield point elongation" to establish a method for choosing the average values for both of these properties as was done in our paper. This method appears to us to give a value which is best suited to the observed phenomena.

The changes taking place at the yield point can in many cases be followed with the eye if great care is exercised. For example, when the load-deformation curve shows an upper yield point, the surface marking accompanying the first stretcher strain remains clearly visible during the time the specimen is being pulled through the yield point elongation. When the whole gage length has been affected, the load-deformation curve rises, as has been described, but it is not until a load corresponding to the load obtained at the upper yield point is reached that the remnant of the first surface marking disappears.

The mechanism of stretcher strain formation just outlined serves to answer, in part, some of the questions raised by Mr. Kenyon as to the influence of the testing machine. In any moving mechanism, there is certain to be some of the effects of inertia. There may be, therefore, a slight "overshooting" of the weighing mechanism which would cause a raising of the load measured by the testing machine at which yielding occurred and temporarily a too great a drop in the load at the beginning of the yield point elongation. It is because of the latter observation that we cannot agree with Mr. Kenyon that the "lower" yield point is of any particular significance if he accepts Bach's definition of "lower" yield point as being the "lowest value to which the stress falls during the yield point elongation." Our reasons for choosing an average value have already been described, but we would like to point out that, in our opinion, the load at which the larger part of the yield point elongation occurs is most significant of the inherent properties of the steel at a given head speed of the testing machine. It is this load which, we believe, should be chosen as the "lower" yield point. As has been shown, the effect of speed of testing on this portion of the curve is so great that when comparing the properties of two different steels it is clearly necessary to use the same head speed of the testing machine. We believe, however, that the effects of the inertia of the weighing mechanism and the elasticity of the different parts of the testing machine on the shape of the load-deformation curve are very small, particularly when a machine is used whose capacity is considerably greater than that of the specimen. To remove all of the effects of the elastically deformed parts of the system would be quite impossible even theoretically. For example, the parts of the specimen which, during the yield point elongation are only deformed elastically, act as a spring and the influence of these parts on those which are being deformed plastically must be considered in any detailed study of the phenomena. Certain it is, in our judgment, that most of the observed phenomena occurring at the yield point of annealed low carbon steel are not due to the method of testing as has been claimed by some authors and to which reference

has been made by Mr. Kenyon. A simple proof of this is that no such phenomena occurs in the load-deformation curves of copper, or nickel, or aluminum when these metals are plastically deformed under exactly the same conditions. The excellent reasoning of Dr. Rawdon and the conception of a network of "unfavorably" orientated grains appeals to us, but the difficulty seems to us to lie in the questions: Does not the same condition exist in aluminum or copper? And, if the same condition does exist, why do not these metals show stretcher strains in deep drawing?

Our experiments and observations indicate, as was pointed out in our paper, that the presence of more than one stretcher strain at a given instant lowers the load at which the yield point occurs and decreases the percentage yield point elongation. In answer to Dr. Krivobok's and Mr. Kenyon's questions regarding this, the reason is because of the reduced rate of deformation. When one stretcher strain is present, there are two areas undergoing deformation and when two stretcher strains are present there are four such areas. Since the head speed of the testing machine is constant, the rate of deformation (in inches per inch per minute) is, then, in the latter case half that of the former. It should be pointed out, however, that the addition of one stretcher strain does not decrease the rate of deformation in the same proportion if more than one stretcher strain is present. It will be noted that the curves in the paper were plotted on a logarithmic scale and so the effect of any additional stretcher strains is small. The difference in scale in Figs. 14 and 15 probably mislead Dr. Krivobok to conclude that the effect of speed is greater than the effect of grain size. This is not true because a fifty per cent increase in grain size appears to make a difference in yield point elongation of about 16 per cent, whereas a fifty per cent increase in the rate of deformation increases the yield point elongation by only some 4 per cent. We agree with Mr. Kenyon that within the limits of practicability the removal of stretcher strains cannot be obtained by adjusting the speed of the press. One of the reasons for this is because the number of stretcher strains present at any given instant in a large stamping is so indeterminate not only between one shape of stamping and another, but also between two stampings of the same shape, that the rate of strain might vary also within extremely wide limits.

In this discussion, we have so far been considering only those parts of the specimen which are undergoing or have undergone plastic deformation. A vital significance of the "upper" yield point seems to us to lie also in a consideration of the status of the conditions existing in the parts of the specimen which are not being plastically deformed. Many carefully made tests indicate that throughout a 2-inch gage length the physical properties of the steel vary only to a small degree. That is, at a given speed of test, the yield point varies by only some 2 per cent and yet the yield point can be raised by increasing the rate of deformation by some 60 per cent or more. The extremely pertinent question then arises: "If the load at which substantial plastic deformation is taking place can be raised or lowered within such wide limits by changing the rate of strain, how is it that during such local plastic flow the remaining parts (which are not undergoing plastic deformation and which are, therefore, not moving) can support loads which they could not support under other conditions? To state this differently, as soon as substantial plastic deformation occurs, the parts of

the specimen not undergoing plastic deformation are in a position of substantial rest. These regions are, nevertheless, able to support during the yield point elongation a load which is far greater than that which they are apparently able to withstand if plastic deformation were occurring in other parts of the specimen at a lower rate of deformation! The absurdity of such a situation appears to us to be obvious. That the metal which is not undergoing plastic deformation can support such high loads without plastically deforming clearly indicates, it seems to us, that the steel can, without plastic deformation occurring, be subjected to a stress far above the load at which the yield point elongation is occurring. That it finally does yield is, as we have described, due to the presence of adjacent cold work (stretcher strain) and stress concentrations. If there were no "upper" yield point, then, stretcher strains would appear almost simultaneously in all parts of the specimen, and consequently their depth would be so slight as to constitute only a roughening of the surface. As we see it, the plastic deformation of copper and aluminum occurs in this manner, i.e., an almost immediate elongation of the whole gage length under an increasing load. Whatever extremely slight differences there may be in the level of any surface markings, it does not seem to us that these can be called "stretcher strains."

We come now to a brief discussion of the rate of work-strengthening and a brief summary of the stress-strain relationship at the yield point. It appears to us that the load-deformation curves are of the shape shown in Fig. 19a of our paper because of the phenomenon which causes the presence of an "upper" yield point. As has been stated, the metal can support an inordinately high load before plastic flow takes place. According to our view, when plastic flow does take place, the behavior of the metal is governed by the magnitude of the load to which it finds itself subjected and by its resistance to further plastic flow. If, on the other hand, it could be possible to measure the loads actually necessary to produce an elongation less than the yield point elongation, the shape of the load-deformation curve would be quite different. In order to make this clear, we reason as follows: At the load A in Fig. 1a of this discussion, the metal finds itself with a high load upon it, and because of this it deforms at a high speed along the dotted line Aa. This small increment of metal cannot deform further until a load greater than that at a is reached. As the load drops to B, the next two increments (one at each end of the first stretcher strain) deform a smaller amount (Bb) at a lower speed. That is, it appears as if the metal at the instant plastic deformation occurs has the same relatively low resistance to further plastic deformation as copper or aluminum, but because of the inordinately high load to which it finds itself subjected is forced to resist this load as best it can. It is the resistance of the metal to plastic deformation, once this has started, which prevents the load at the yield point elongation from dropping even lower. We believe, then, that the actual stress-strain curve at zero rate of strain is of the shape shown in Fig. 2 of this discussion. The rate of work-strengthening may be defined as: the increase in stress per unit deformation. It is the slope of that portion of the curve shown by LM in Fig. 2 which reveals the rate of work-strengthening. As is evident, the rate of work-strengthening changes as the stress becomes greater. Work-strengthening, as we have pointed out, and as is emphasized by Dr. Sauveur,

cannot take place until after plastic deformation has begun, but, as we see it, the speed of deformation influences the rate of work-strengthening only slightly. As an analogy, the rate of work-hardening of a metal is found by plotting the increase in hardness against the percentage reduction by cold rolling. The shape of this curve is practically the same whether or not the cold rolling is done slowly or rapidly. It requires a greater load to produce a given percentage reduction in a short time than it does to produce the same percentage

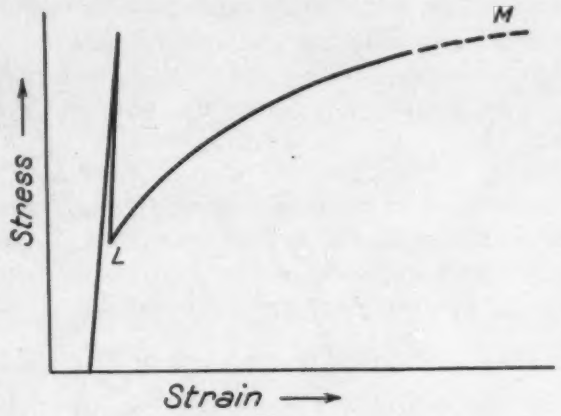

Fig. 2

reduction in a longer time, but the rate of increase in hardness, i.e., the slope of the curve, is only slightly affected, if at all. A point we wish to emphasize in connection with the paragraph on page 182 of our paper is that it is incorrect, in our opinion, to give as the rate of work-strengthening of the metal during the yield point elongation (as some have done) the slope of a straight line joining the stress corresponding to the load at A in Fig. 1a and the stress corresponding to the load a in the same figure. Plastic deformation does not begin until the stress at A has been passed, and it is, therefore, incorrect, it seems to us, to use this stress as the basing point of the rate of work-strengthening curve. We agree with Mr. Kenyon that the conditions on page 184 of our paper could hold for a nonferrous metal and, indeed, we believe that insofar as the rate of work-strengthening is concerned, the plastic deformation of iron and low carbon steel takes place in much the same manner. The important difference in their behavior is that iron and low carbon steel can be subjected to an abnormally high stress before plastic deformation occurs, whereas copper and aluminum yield at a low stress. The initiation of plastic deformation in the nonferrous metals, copper, aluminum, etc., occurs at low stresses and a higher stress is immediately necessary to produce further plastic deformation. In iron and low carbon steel, the initiation of plastic deformation requires an abnormally high stress, but when this stress has been reached and the resistance has been lowered, the steel then behaves, like copper and aluminum, in a perfectly normal manner. The explanation of this behavior of iron and low carbon steel does not lie, then, as we see it, in the answer to the question "why is the metal so weak at the instant plastic deformation occurs," but in the answer to the question, "why is the metal so strong before plastic deformation occurs?"

COLD WORKING OF HOLLOW CYLINDERS BY AUTO-FRETTAGE

By Norman E. Woldman

Abstract

This paper, in the present Symposium on the Plastic Working of Metals, deals only with the theory and practice in the cold working of monoblock hollow cylinders or tubes by radial expansion methods, known as "auto-frettage." The writer does not discuss or develop the mathematical theory of the elastic strength of auto-frettaged cylinders, nor to include design formulas and calculations, since they are so thoroughly considered and so well written in detail in the references cited. The theory, practice, types and applications of equipment, and metallurgical aspects of this process are herewith outlined and discussed.

METALS may be either hot-worked or cold-worked plastically. These processes involve (1) forging by the hammer, drop, press or upsetter, (2) rolling, (3) piercing, (4) extruding, (5) drawing and (6) radially expanding or "auto-frettage." Usually, the dividing line between hot and cold working is the critical temperature or recrystallization temperature of the metal or alloy being worked. If the metal or alloy is worked above this temperature, it is said to be hot-worked; while if the metal or alloy is worked below this temperature, it is said to be cold-worked. It is evident therefore that cold working may take place at atmospheric temperature or at some limited elevated temperature.

Cold working results in hardening of the metal, sometimes referred to as "work-hardening" or "strain-hardening." This involves deformation and fragmentation of the grains of which the substance is built up. The deformation is parallel to the direction of the tensile stress and perpendicular to the direction of the compressive stress. Slip occurs on the gliding planes of individual crystal grains, accompanied by the formation of thin layers of nonelastic material which is capable of setting into a hard, nonyielding condition. This deformation and fragmentation of the grains eventually become con-

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siderable along these cleavage or gliding planes, resulting in increased hardness and strength with subsequent decrease in plasticity. The interference to further slip and deformation along this hardened and strengthened boundary causes the occurring stresses to be directed in a different direction or along an adjacent crystal or cleavage plane rendering the different crystal grain boundaries harder, stronger and more resistant to deformation. In plastic deformation relative motion is confined to many block movements along slip planes, the crystalline fragments between such planes moving as units in which the atoms retain substantially their initial relative positions with respect to their neighbors. However, the thin layers between these blocks constitute the material which has been deformed. Grain deformations are permanent when effected below the annealing or recrystallization temperature.

The radial expansion or "auto-frettage" method of cold working steel takes place at atmospheric temperature. This process, which is chiefly applied to cylinders and tubes of heavy wall thickness, is a method by means of which the bore of the cylinder or tube is cold-worked through the application of a radial pressure on the bore. There are several ways of applying this radial pressure, but hydraulic methods have proven most successful. "Auto-frettage" may be defined as the process of automatically setting up the effects of shrinking a number of infinitely thin tubes or hoops over one another to make up the wall thickness of a cylinder or tube in such a manner that the inner layers are in states of residual compression and the outer layers in states of residual tension. These effects are produced in a monoblock cylinder or tube by the application and release of a radial pressure on the bore, which pressure, during its application, sets up an overstrain in some or all of the layers in the wall. In other words, the condition of "auto-frettage" is produced by the application and release of a high liquid pressure in the bore, which in turn sets up a positive permanent deformation in some or all of the layers in the metal walls of the cylinder or tube. The internal hydraulic pressure expands the cylinder until practically all the metal has been stressed beyond its elastic limit. The metal at the bore is deformed the most, and having yielded plastically, attempts to maintain its permanent set, but those outer layers, which have been stressed within the elastic range, attempt to return to their original dimensions, thereby compressing the inner layers. The outer layers retain part of their tangential tension, while the inner layers take on a tan-

gential compression. The greatest compression is at the bore, the greatest tension is at the outer surface, and somewhere near the middle of the wall, there is metal free from tangential stress.

Theory of "Auto-Frettage"

From observed behavior of steel specimens under tension, compression and shearing stresses, and from the effect of combinations of these stresses on the specimens, theories and formulas for radial expansion were developed. Captain Duguet (1)[1] of the French army determined the relationship between the elastic limits in tension, compression and shear, and he published in 1885 his results in a book entitled "Limite D'Elasticite et Resistance a la Rupture." Later, in 1912, Pierre Malaval (2) published a set of formulas for the resistance of cylinder tubes under radial expansion, based partly on his own theories and experiments, and partly on those of Captain Duguet.

"Auto-frettage" involves the study of elasticity and plasticity of the metal cylinder or tube. Elasticity is considered as the property or ability of a metal to recover its original form on removal of the applied load. If the recovery is complete, the metal is said to be perfectly elastic. However, if the load per unit area exceeds a certain limiting value, which we call the elastic limit of the material, complete recovery is not obtained and a small permanent set or strain is obtained on release of the load. Permanent deformation involves displacement within the metal, that is, sliding of crystal fragments over each other. This type of deformation is called plastic deformation, and the property of plasticity depends upon the ability of these displaced crystal fragments to re-establish themselves or of sustaining permanent deformation without rupture.

Two simple laws of mechanics must be considered, namely, Hooke's Law and Poisson's ratio. Hooke's law states that in small strains of elastic bodies the stress is proportional to the strain, i.e., within the limits of elasticity of a material, the strain set up is proportional to the stress producing it. Poisson's ratio refers to the phenomenon that when a bar is subjected to an axial tensile load, the bar is elongated in the direction of the load, and at the same time the lateral dimension of the bar decreases. In other words, when a

[1]The figures appearing in parentheses refer to the bibliography appended to this paper.

bar is subjected to direct loading it extends longitudinally and contracts transversely, or vice versa, according to the direction of the load. Poisson's ratio is, therefore, the ratio of the transverse strain to the longitudinal strain, and is a constant (about ⅓ for steel) for a given material within the limits of the elastic limit. In bodies which are homogeneous and isotropic there are two principal kinds of elasticity, that in virtue of which the body resists change in volume and that resisting change of shape.

Test pieces (6) which had been stressed beyond their elastic

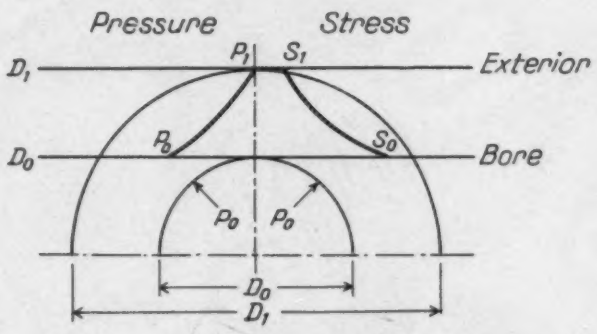

Fig. 1—Diagram Representing the Character of the Pressures and Stresses in the Wall Thickness of a Monoblock Steel Cylinder.

limit show on reloading increasing extensions with equal increments of load. Hooke's law is no longer obeyed even at low loads. The elasticity of the steel, however, recovers slowly at ordinary temperature and more quickly at higher temperatures. After complete recovery, the steel again obeys Hooke's law up to its new elastic limit, which is always higher than the original elastic limit and may be higher than the load applied to overstrain the steel. Similar behavior is observed in compression after overstraining in compression. Howe (3) concluded that the effect of overstrain occurring with repeated reversals of stress is anisotropic, i.e., the overstrain strengthens the metal against later stresses in the same direction and weakens it against stress in the opposite direction. The effect of simple overstrain, however, is at least partially isotropic, i.e., the metal is strengthened in all directions, but to a degree which varies with the direction of the later stresses, being greatest for stresses occurring in the direction of the overstrain itself.

A radial pressure applied to the bore of a cylinder or tube will set up pressure and stresses in the wall thickness which can be easily calculated. Fig. 1 represents the character of the pressures and

stresses in the wall thickness of a monoblock steel cylinder. Let D_1 and D_0 represent the exterior and bore diameters of the cylinder. When a radial pressure P_0 is applied to the bore, pressures and stresses are set up in the wall thickness which may be represented by the curves P_0 P_1 and S_0 S_1 respectively. For a class of steel of uniform elasticity, the bore of the cylinder will not be overstressed by the applied pressure P_0 if S_0 does not exceed the tensile elastic

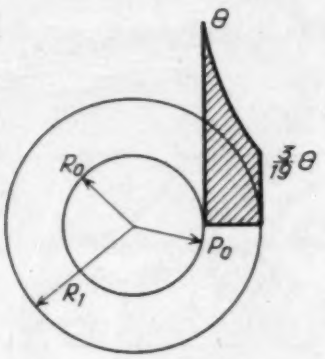

Fig. 2 — Distribution of Stress in the Cross Section of a Cylinder.

limit of the steel. If the stress is less or equal to the elastic limit, the cylinder will be stressed near or to its maximum transverse or radial elastic resistance by the bore pressure. Repeated applications of radial bore pressures up to the limiting value will stress and strain the steel in the wall within its elastic range, and, on release of such pressures, the cylinder wall will return to its original dimensions.

The stresses within the walls of the cylinder are not uniformly distributed. The application of an internal radial pressure produces stresses in three directions, namely, radial, longitudinal and tangential. As the tangential stresses are the greatest they are the limiting factors in the elastic strength of cylinders. These tangential (18) stresses are tensile stresses which vary from a maximum at the interior of the bore to a minimum at the exterior of the cylinder as shown in Fig. 2, where the actual intensity of stress at any radius is plotted as ordinates.

For a cylinder of given diameters, the safe or maximum radial bore pressure is limited to a certain value, that is, when the tensile elastic limit may be reached at the bore. The greatest pressure which can be confined in such a cylinder without stressing the inner layers beyond the elastic limit, is only 63 per cent of the elastic limit of the

steel from which the cylinder is made. The stress in the outer layers of the cylinder having a wall thickness equal to its interior diameter is only 3/19 of the stress at the inner layer.

The elastic limit at the bore must not be exceeded if the cylinder is to function elastically, that is, return to its original dimensions after the applied pressure is released. However, there are several methods which may be utilized to increase the elastic strength or

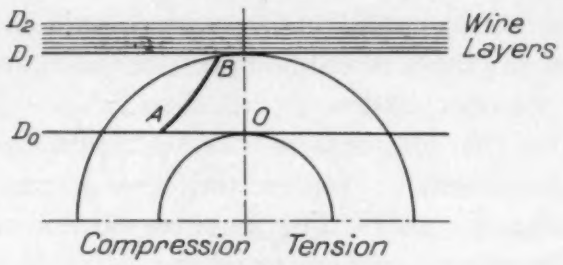

Fig. 3—Diagram Showing the Compressive Stresses Set Up by the Wire Wound at Suitable Tensions Around the Exterior Diameter.

radial elastic resistance of the cylinder. These methods are (1) increasing the wall thickness, (2) using a steel of higher elastic limit and (3) inducing an initial compression in the bore. Increasing the thickness of the wall beyond a certain value does not materially increase the elastic strength of the cylinder. Langenberg (10) showed that very little effect upon the elastic strength of steel is obtained by increasing the wall thickness beyond the bore diameter. The greatly increased weight brought about by increasing the wall thickness beyond a 3 to 1 diameter ratio is distinctly out of proportion to any increase in elastic strength. However, below a 3 to 1 diameter ratio, a marked increase in elastic strength is obtained with an increase in diameter ratio. Steels of higher elastic limits are to be used with caution owing to the lower ductility and impact values usually associated with high elastic limits. Inducing an initial compression in the bore to increase the elastic strength involves the principle of "auto-frettage" whereby the bore will be under a compressive stress when the initial bore pressure is zero. The internal pressure will then have to overcome this compressive stress before the steel will be stressed in tension.

Initial compression in the bore of a cylinder may be obtained by several methods. The cylinder may be compressed by winding wire at suitable tensions on the exterior until the required initial tensions

are obtained, or by shrinking an outer tube or jacket on the cylinder, or by radial expansion of the bore. Introducing initial compression by radial expansion of the bore can be performed by forcing tapered mandrels through the bore, or by explosive effects (powder pressure), or by hydraulic pressure. Langenberg (10) cites a case where General Rodman, about 1850, obtained an ultimate compression of the metal adjacent to the bore and an increased elastic strength by cooling a cast cylinder in such a manner (water-cooling the core of the casting) that the interior would chill first. Colonel Dickson (16) described a method to obtain initial compression in the bore by centrifugal casting of a steel cylinder.

Fig. 3 shows the compressive stresses (curve AB) set up by the wire wound at suitable (either uniform or variable) tensions around the exterior diameter. The inner, or bore, diameter is represented by D_o, the exterior diameter of the cylinder by D_1, and the layers of wire by D_1D_2. Although the cylinder wall is in compression, the wire will be in residual tension.

Fig. 4 shows the compressive stresses (curve AB) set up by shrinking one tube over another, that is, by shrinking an outer tube or jacket over the cylinder. D_o and D_1 represent the bore and ex-

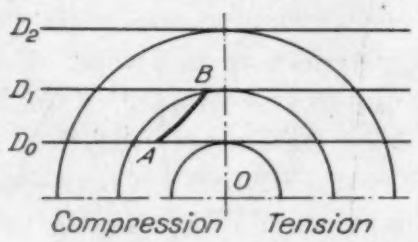

Fig. 4—Diagram Showing the Compressive Stresses Set Up by Shrinking One Tube Over Another.

terior diameters respectively of the inner cylinder, while D_1 and D_2 represent the bore and exterior diameters respectively of the outer tube or jacket. This process is performed by machining the interior diameter or bore of the outer tube a little smaller than the exterior diameter of the inner cylinder. From predetermined calculations the outer tube is heated until it expands sufficiently to fit over the inner cylinder. The outer tube is then shrunk onto the inner cylinder by cooling. And, in attempting to shrink back to its original diameters, the jacket compresses the inner cylinder setting up compressive stresses as represented by curve AB. Here again, as in the case of the wire-wound cylinder, the outer jacket will be in residual

tension. In addition, the inner tube, owing to its compressive elasticity, exerts a pressure on the bore of the outer tube. The pressure is radial and acts upon both tubes at the surface of contact. The resultant diameter of the surface of contact is between the original diameter of exterior of the inner cylinder and the bore of the outer tube. Several tubes can be shrunk together so that each tube will do a share of the work resisting the stresses set up by a radial pressure on the bore. Malaval (2) and Macrae (14) showed that the ideal would be reached if each tube were worked to its elastic limit, that is, worked to its maximum without fear of any permanent expansion. Under these conditions the greater the number of tubes which make up the wall thickness of the cylinder, the greater will be its radial strength or transverse resistance. Therefore, a cylinder made up of a number of infinitely thin tubes correctly shrunk together would be a cylinder of maximum resistance. A compound cylinder (10) of two layers (3 to 1 diameter ratio) will have an elastic strength of about 0.86 times the elastic limit of the steel from which the cylinders are made. However, such compound cylinder made of an infinite number of layers will have an elastic strength of 1.26 times the elastic limit of the steel.

From the above it is evident that a combination of the wire winding method and the tube shrinking method can be used to obtain initial compression in an inner tube. The range of elasticity in the inner tube, that is, original tensile elastic limit plus initial compression, produced by these methods, enable a pressure greater than the radial bore pressure to be applied without exceeding the tensile elastic limit at the bore.

Referring to Fig. 1, it is possible to consider the wall of the cylinder constructed of an infinity of cylinders infinitesimally thin, fitted one within another with a proper shrinkage. The wall thickness between the bore and exterior diameters may, therefore, be considered to be made up of an infinite number of layers of steel. If the internal pressure is increased until the stress in the bore layer is less or equal to the tensile elastic limit of the steel, the bore layer and the outer layers will be stressed within the elastic range of the steel. However, if the bore pressure is increased until the stress in the inner layer just exceeds the tensile elastic limit of the steel, the bore layer will be overstressed and given a positive permanent set, while the outer layers will be stressed within the elastic range of the steel. Further increase of the bore pressure will overstress the layer next

to the bore layer, increasing the magnitude of the permanent set in the bore layer and the elastic stresses in the remaining layers. It becomes evident, therefore, that, as the bore pressure continues to increase, the layers further away from the bore layer will in sequence be overstressed and the permanent sets of the inner layers further increased. Each layer of steel from the bore to the exterior will behave according to the stress-strain relationship of the class of steel

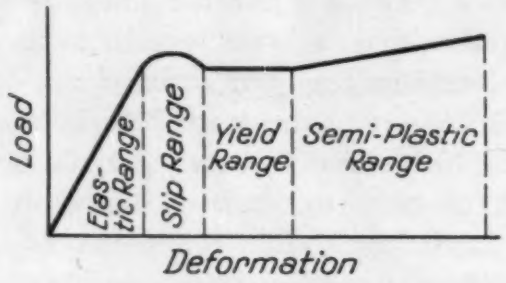

Fig. 5 — Diagram Representing the Stress-Strain Relationship of the Steel During Overtension.

used, that is, as the radial bore pressure increases, the bore layer will in turn be stressed to the elastic range, through the yield range and then into the semi-plastic range. The layer next to the bore will in its turn be stressed in these ranges, but here the permanent deformation will be less than in the bore layer. Layers in sequence more remote from the bore will be overstressed to gradual lesser degrees, until a layer is reached where the set-up stress is just equal to the elastic limit of the steel. Layers more remote from the bore than this particular layer will be stressed in the elastic range of the steel.

Macrae (14) showed these stresses (elastic, yield and semi-plastic) may be represented by curves in the following diagrams. Fig. 5 represents the stress-strain relationship of the steel during overtension.

Fig. 6 shows a monoblock cylinder with the steel stressed in three zones, the value of stresses being represented by S_0 S_1 S_2 S_3.

"On application of the radial bore pressure, the inner layers from the bore to a diameter represented by D_2 are overtensioned. The deformations set up in the layers from bore diameter D_0 are elastic. On release of the bore pressure, layers which were in states of elastic stress (diameters D_3 to D_2) endeavor to return elastically to their original dimensions. Layers in which the set up deformations were both elastic and permanent (diameters D_2 to D_0) are relieved of most of their elastic deformation, but owing to the perma-

nent deformations set up by the bore pressure, are unable to return to their original dimensions. Now these layers (D_2 to D_0) which were overstressed on application of the radial bore pressure have lost most of their compressive elasticity, and behave as a plastic metal to compressive stresses. The outer layers (D_3 to D_2) in endeavoring to return to their original dimensions exert compressive stresses on the plastic layers, set up compressive deformations in these layers,

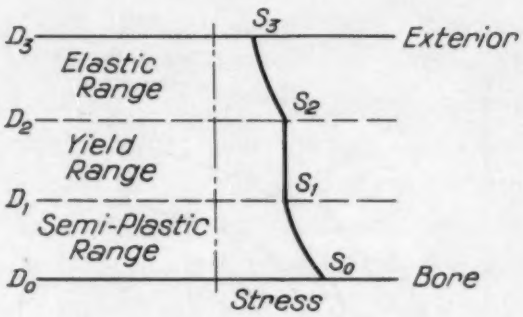

Fig. 6—Diagram Showing a Monoblock Cylinder with the Steel Stressed in Three Zones.

but by reason of the resistance offered by the plastic layers remain in states of residual tension. The permanent deformation set up in the bore layer on application of the radial bore pressure is greater than that set up in the second layer. By reason of this difference, on release of pressure the bore layer offers a resistance to the closing in of the second layer. In these two bore layers, therefore, the effects of shrinking one layer on the other are automatically obtained; relatively, the inner layer is in a state of compression and the outer layer is in a state of tension. The third layer from the bore sets up additional compressive stresses in the two inner layers and in turn tends to remain in a state of relative residual tension. The effects of shrinking an outer layer on two shrunk up layers are produced. Throughout the overtensioned layers, therefore, the effects of continuous shrinkage are obtained. In the cylinder under consideration the release of the bore pressure produces the effects of "continuous shrinkage," "self-hooping" or "auto-frettage" in the wall thickness of the cylinder; that is, the bore layer and some of the inner layers will be in states of residual compression, while the remaining outer layers will be in states of residual tension."

Auto-frettaged monoblock cylinders are unstable and must be aged or given a low temperature heat treatment to produce stability, that is, to make elastic for repeated applications of pressure up to and

including the auto-frettaged pressure. Cold working of the bore lowers the tensile elastic limit and reduces to practically zero the compressive elastic limit. However, if this overtension, due to cold working, is followed by a low temperature anneal the steel recovers and the tensile elastic limit increases to a value equal to or greater than the applied auto-frettage pressure. The net effect of the low temperature anneal varies in the three different zones. In the outer or elastic zone,

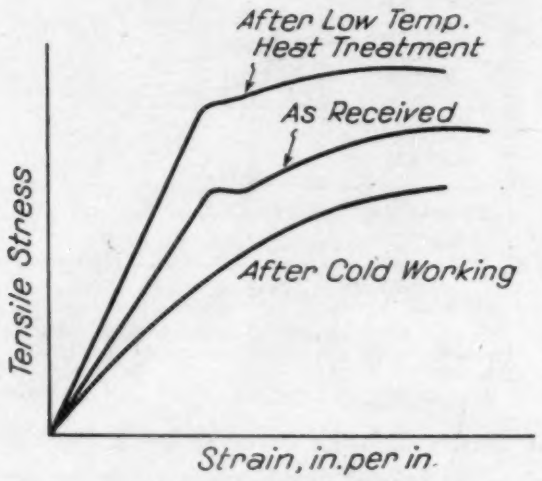

Fig. 7—Diagram Representing the Character of the Stress-Strain Curves for the Monoblock Cylinder Before Radial Expansion, After Cold Working, and After the Low Temperature Heat Treatment.

where the layers are under residual tensile stress, the low temperature heat treatment will increase the tensile elastic limit; and in the inner or semi-plastic zone, where the layers are under residual compressive stress, the low temperature heat treatment will increase the tensile and compressive elastic limits; while in the middle or yield zone, the low temperature heat treatment will increase the tensile elastic limit and decrease proportionately the compressive elastic limit. The greater the original auto-frettage pressure the greater the increase in elastic limit or elastic strength after recovery by low temperature heat treatment.

The rate of recovery (8) of elasticity after simple overstrain in tension varies greatly with different steels, the harder steels in general recovering more slowly. The greater the degree of overstrain, the longer the time required to bring about recovery at a given temperature, or the higher the temperature required to bring about recovery in a given time. All steels show some progress towards recovery at atmospheric temperature. The rate of recovery depends

on the composition of the steel, the heat treatment given to the steel prior to cold working, the magnitude and rate of overstraining during cold working, and the temperature and time of low temperature heat treatment.

Recovery of elasticity on aging or low temperature heat treatment is accompanied by increase in hardness, increase in tensile strength, increase in elastic strength, decrease in plasticity, decrease in per cent elongation, decrease in per cent reduction in area and decrease in impact value. The latter two properties are less affected.

The acquired elastic strength was found to be about 98 per cent of the original expanding pressure during "auto-frettage."

The character of the stress-strain curves for the monoblock cylinder before radial expansion, after cold working, and after the low temperature heat treatment, may be represented in simple manner by Fig. 7. The curves are self-explanatory in that they show the decrease in the elastic limit after cold working, and the recovery and increase in the elastic limit after low temperature heat treatment.

Equipment, Methods, and Practice

It was previously stated that initial compression in the bore of a monoblock cylinder may be obtained by winding wire at suitable tensions on the exterior of the cylinder, by shrinking an outer tube over the cylinder or by radial expansion of the bore. It was further stated that radial expansion (auto-frettage) may be performed by forcing a mandrel through the bore, by explosive (powder pressure) effects, or by hydraulic pressure. As the only efficient and successful method of "auto-frettage" is by hydraulic pressure, this practice will be herewith outlined. For further information concerning the other methods the reader is referred to the literature and references cited.

There are at present two methods of "auto-frettage" by hydraulic pressure used effectively, namely, (1) The "Open Method" which has no exterior constraints to limit the amount of expansion of the monoblock cylinder, and (2) the "Container Method" in which the cylinder is placed inside a strong thick-walled, hollow tube container that limits the expansion of the cylinder to the desired amount. The latter has the advantage over the former in that monoblock cylinders, having sections of different wall thickness or having a tapered contour, may be cold-worked by a single operation using a bore pressure sufficiently high to cold-work the thickest-walled section the

desired amount; while in the former method each section has to be cold-worked separately with bore pressures proportional to the wall thickness in each section. The disadvantage of the Container Method is that it covers up the existence of soft spots and irregularities in the structure condition of the metal which can be detected by the Open Method.

The essential apparatus necessary for the "auto-frettage" are:

(a) A low pressure water pump to operate the intensifier.

(b) An intensifier to develop high hydraulic pressure.

(c) An arbor or spool to be secured in position in the bore of the cylinder.

(d) High pressure pipes and connections to lead high pressure liquid from the intensifier to the cylinder.

(e) Packings to hold the high pressures.

(f) Pressure gages and accessories to measure the pressures developed.

(g) Star gages to measure the bore diameters before and after cold working.

(h) Micrometers or indicators to measure the outside diameters of the cylinder during expansion.

(i) Liquid medium of small compressibility under pressure, such as glycerine, to be pumped into the bore of the cylinder to develop the internal pressure.

(j) The monoblock cylinder to be cold-worked.

(k) A container if the container method is to be used.

Patents held by A. H. Emery (5) described equipment and process designed to produce the interior pressure by forcing into the breech end of a battleship gun, through a stuffing box packing, a cylindrical steel piston compressing the special liquid within the bore of the gun. The bore of the gun was sealed against the high internal pressure by two fixed packings at the breech and muzzle ends. The expanded steel was stabilized by introducing live steam at a pressure of 15 pounds at a temperature of 250 degrees Fahr.

Langenberg (10) described an experimental unit to expand a cylinder having an eight-inch bore and an eight-inch wall thickness.

Low pressure water was delivered by a pump through the bottom of the intensifier (Fig. 8) compressing the high pressure liquid in the interior chamber of the piston. This high pressure liquid produced by the intensifier was led into the bore of the cylinder through a spool which was fitted into the cylinder as shown in Fig. 9. The

packing to hold the pressure at each end of the spool consisted of a series of concentric metal and rubber rings. Several dial gage holders were placed around the cylinder and at each end to measure the expansion of the exterior and the changes in the length as the pressure increased. The high pressure liquid was forced into the annulus between the inner diameter of the cylinder and the exterior diameter of the spool.

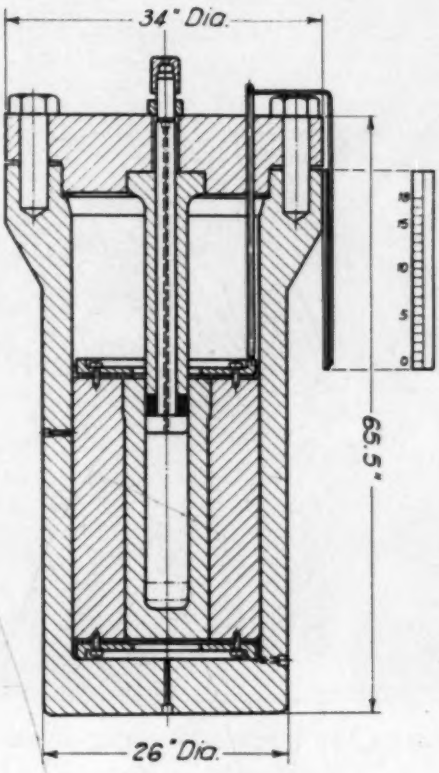

Fig. 8—Sectional View of Hydraulic Intensifier Used in Auto-Frettage Work. The Ratio of Intensification in This Apparatus is 19.7 to 1.

Fig. 10 is a photograph of a gun manufactured by the cold working process at Watertown Arsenal (10). The container used to limit the amount of enlargement is also shown. The container is so constructed and its interior contour so machined that each section of the gun of different wall thickness will receive the proper enlargement. The walls of the container limit the expansion. The actual manufacture of the gun by the cold working process is shown in Fig. 11. The container is erected in a vertical position, the gun being lowered into place. Packings are provided at each end of the

gun, the pressure being applied through the pump and intensifier previously described. The end pressure on the packings is held by tie rods. To determine the pressure at which the various sections of the gun make contact with the container, small holes are drilled through the walls of the container in which dial gages with extension points are placed, the gages being assembled in frames.

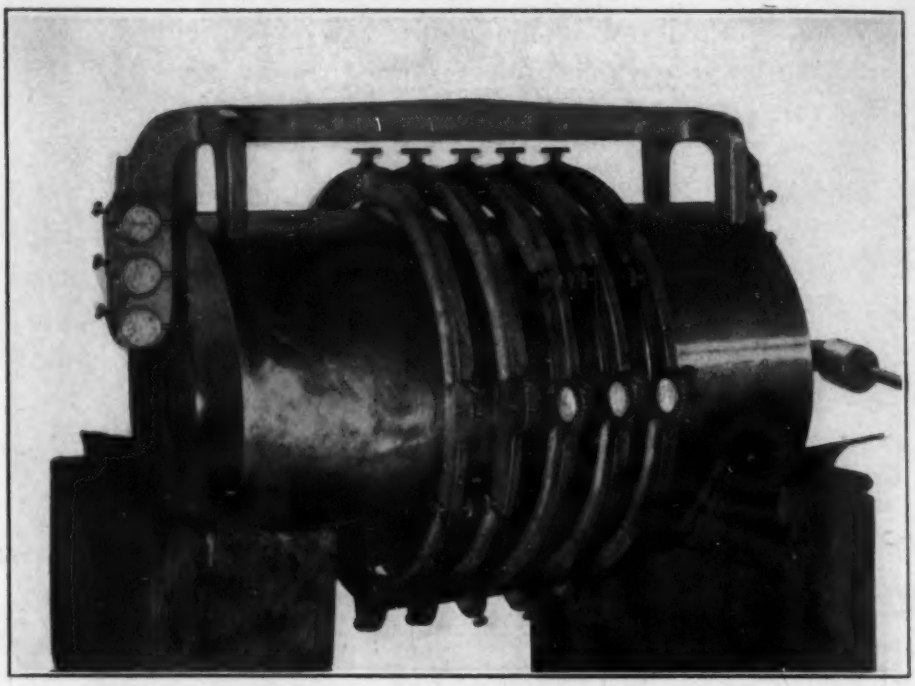

Fig. 9—Cylinder Having an 8-Inch Bore and an 8-Inch Wall Thickness Assembled for Test.

Macrae (14) described an experimental unit in which the steel cylinder A (Fig. 12) was erected in a hydraulic press B which received its hydraulic pressure from an intensifier on the main general supply C. The ram E of the press on being pushed up lifts the cylinder and compresses the liquid G against the top plunger F. The liquid under pressure in the cylinder is forced through a small hole H near the bottom of the cylinder into the high pressure gage J. The bore of the cylinder is closed by top and bottom plungers, the bottom plunger K remaining in position while the cylinder is forced over the top plunger F, both plungers being fitted with packing L.

Macrae described the plant used in the Royal Gun Factory, Royal Arsenal, England, for manufacture of gun bodies by auto-frettage. (Fig. 13) The hydraulic press is fitted with a push-up ram, the normal pressure being 800 pounds per square inch and acting on the

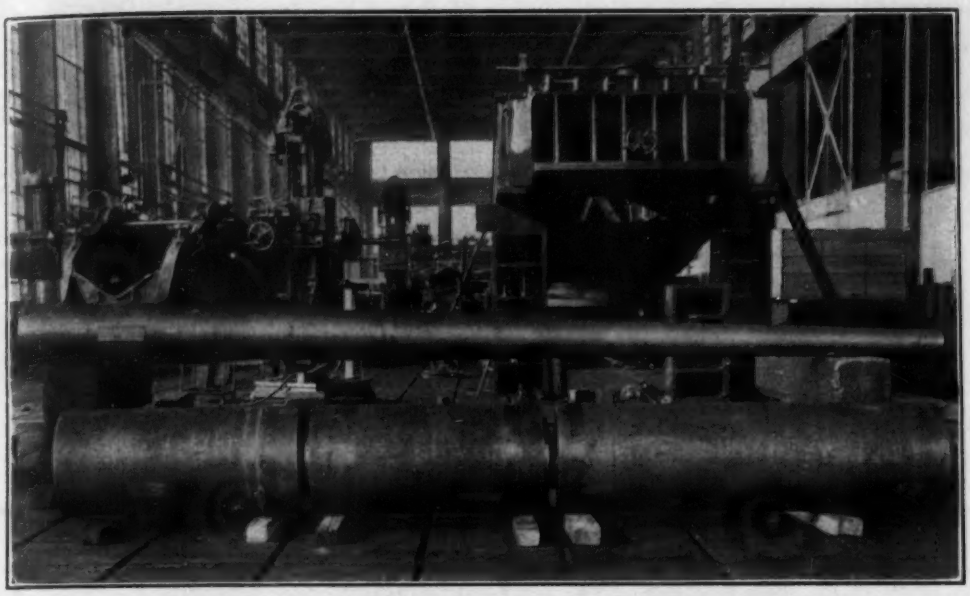

Fig. 10—Photograph of a Gun Manufactured by the Cold Working Process. The Container Used to Limit the Amount of Enlargement is Also Shown.

bottom of the ram. When this pressure has been reached, the power is diverted by means of an automatic ram through a 12 to 1 intensifier which boosts up the pressure to between 6000 to 8000 pounds per square inch. On top of the ram step-up intensifiers are erected to develop a liquid pressure sufficient for the auto-frettage of the gun. The forging is fitted with its core bar, end packings, and screw plugs. supported on Y-shaped supports and jacks, and placed in position for connecting the breech-end of the forging to the step-up intensifiers in the hydraulic press. The intensifier is filled with glycerine until it overflows into the annulus of the arbor, then the liquid is forced into the annulus surrounding the core-bar in the gun by means of a hand pump until it overflows from the escape hole. The intensifier is hand filled to the top of the inlet hole and then coupled to a hand pump. The liquid pressures are applied and measured. The exterior expansion of the forging under load is also measured. The lift of the ram of the hydraulic press, controlled by the operator, forces up the plunger in the intensifier, and in compressing the cylinder sets up a pressure throughout the system. These pressures are recorded on the high pressure gage. After removal of the pressure, the packings, plugs and core-bar are removed, and the forging is suspended horizontally in a tank of cold oil for low temperature treatment. The oil is heated to 250 degrees Cent. and held at that

temperature for a definite length of time. The forging is then removed, cleaned and cooled in air.

Heavy-walled monoblock cylinders of uniform exterior and bore diameters along the entire length may be auto-frettaged through its whole length in one loading. But, with a tapered outside surface

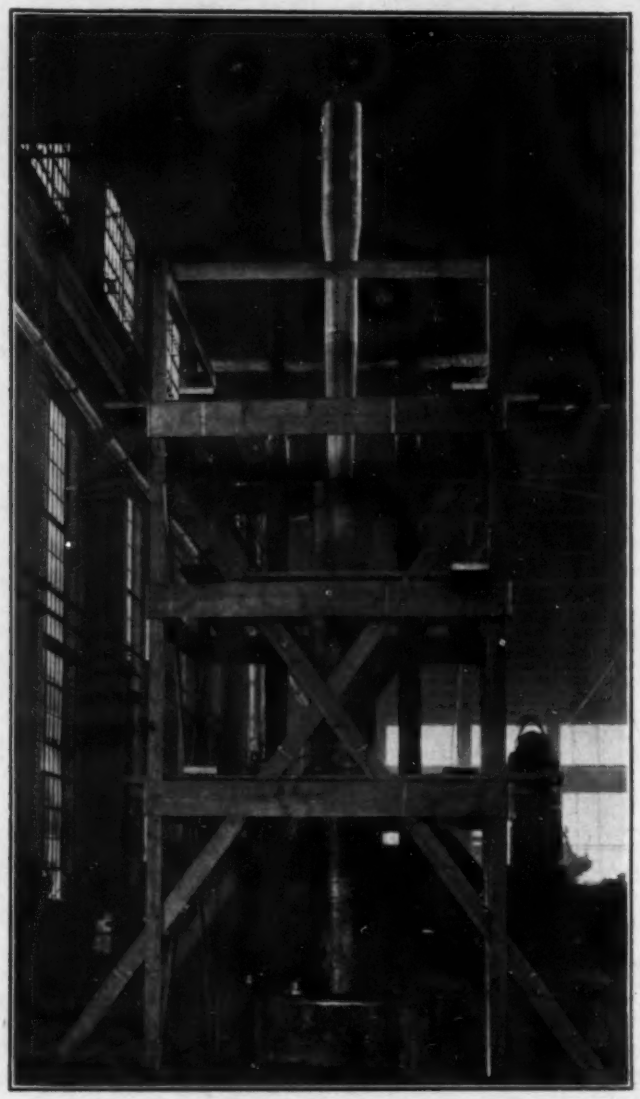

Fig. 11—Photograph Showing the Manner in Which the Actual Manufacture of Guns by the Cold Working Process is Conducted.

and a uniform bore diameter, there are, with uniform internal pressure, differences in the stressing of the metal at different points along the length of the cylinder. Therefore, it becomes necessary to expand one portion of the length, release the pressure, then expand

a second length unit, and so on for all the sections of different wall thickness until the cylinder is expanded along its whole length.

Cylinders, after expansion, usually have to be turned down and rebored to finished dimensions. This removal of metal from the exterior and the bore results in a decrease in the new elastic strength developed after expansion and low temperature treatment. Macrae (14) stated that on turning down only, the radial bore pressure, for

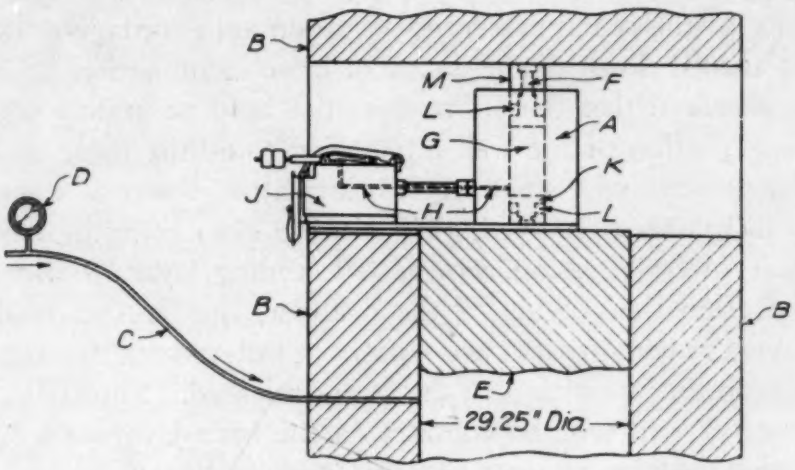

Fig. 12—Sketch of Macrae Experimental Unit.

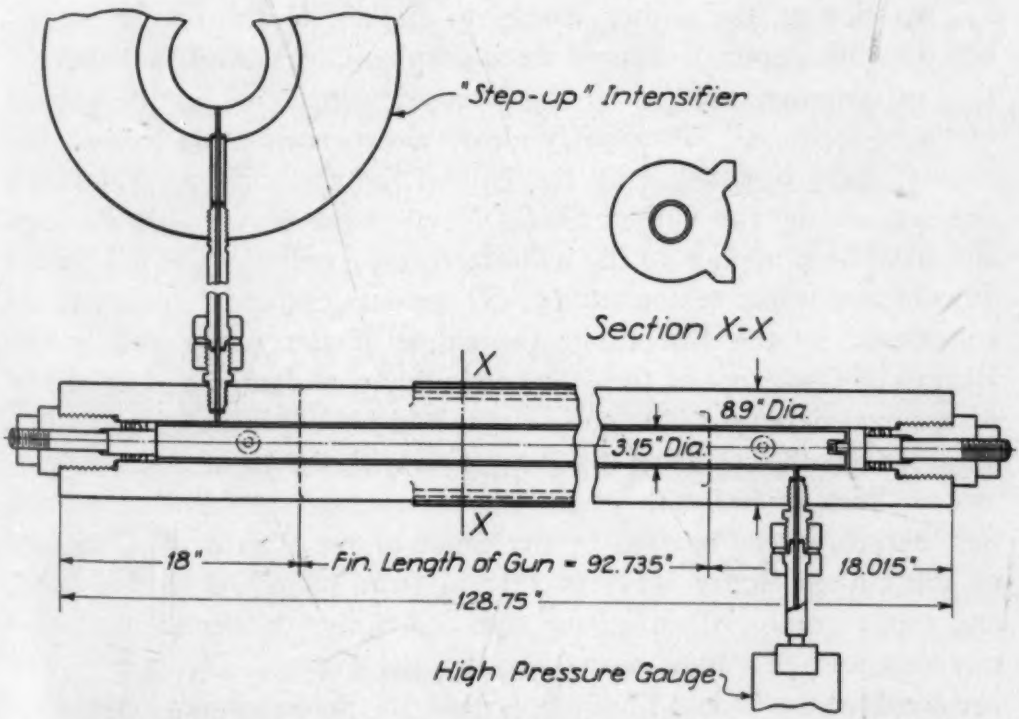

Fig. 13—Sketch Showing General Arrangement of Forging in Position for Auto-Frettage.

which the turned down cylinder will be stable, is equal to the auto-frettaged pressure of the original cylinder less the radial pressure set up on auto-frettage at the diameter to which the cylinder is turned down. On boring only, the radial bore pressure, for which the bored out cylinder is stable, will be that radial bore pressure set up on auto-frettage at the diameter to which the cylinder is bored. And for combined turning down and boring out, the radial bore pressure, for which the machined cylinder is stable, is equal to the difference of the radial pressures set up on auto-frettage at the bored out and turned down diameters. On close examination, however, it will be observed that the above does not hold accurately due to the remaining portion of the wall attempting to adjust itself to the new conditions and to seek stable equilibrium after a part of the outer or inner wall has been removed. The outside (18) being in tension and the inside in compression, as each succeeding layer of the metal is removed from the outside of the cylinder, the stresses holding the inner layers in compression are somewhat relieved and the bore diameters enlarge in proportion to the stress released. Similarly the outside diameter will tend to contract as the bore layers are removed. These machined layers may be considered as hoops, whose removal reduces the final elastic strength of the monoblock cylinder.

In closing, the author wishes to explain to the reader the absence in this paper of desired data, graphs, illustrations and descriptive information relative to the latest developments in the subject of "auto-frettage." Practically all developments in this field in this country have been made by the United States Army at Watertown Arsenal and by the United States Navy at the Naval Gun Factory, and have been applied to the manufacture of radial expanded service guns of major and minor caliber. These developments, therefore, are considered by the Bureau of Ordnance of the Army and by the Bureau of Ordnance of the Navy as highly confidential and as secrets of national defense. The author obtained his experience and knowledge of the subject during his association with the United States Naval Gun Factory and with one of the contracting steel companies, and has therefore been advised by the Chief of the Bureau of Ordnance of the United States Navy to refrain from including in this paper any data, graphs, illustrations and descriptive information of the methods and processes as applied by the Navy. This paper was reviewed, censored and finally approved for publication, after several recommended changes and deletions, by the above Bureaus.

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DISCUSSION

Written Discussion: By H. C. Mann, Senior Materials Engineer, Watertown Arsenal, Watertown, Mass.

In view of recent advances in ideas concerning the true phenomena of plasticity of metals, the theories advanced in the first part of the paper are considered superseded by the newer and better established ones. An excellent survey of the entire field is found in the work of Prof. P. W. Bridgeman, "The Physics of High Pressure," and in the book "Distortion of Metal Crystals" by C. F. Elam, Oxford 1935, with the best theory and one which has the support of several leading authorities, presented in "The Mechanism of Plastic Deformation of Crystals" by G. I. Taylor, *Proceedings*, Royal Society, Vol. 145, No. A855, 1934.

The author's diagram (Fig. 7) and accompanying discussion intended to depict the characteristic behavior of material before cold work, after cold work, and after stabilizing, are misleading. In the diagram the "After Cold Working" curve should extend upward almost to the yield position of the "After Low Temperature" curve. This latter curve should be on the same modulus line as the "As Received" curve since there is no change in the modulus of elasticity produced by cold working.

In his discussion of the return of elastic strength by low temperature heat treatment the author appears to consider the terms "proportional" and "elastic" strength as applying to the same condition, where as a matter of fact there

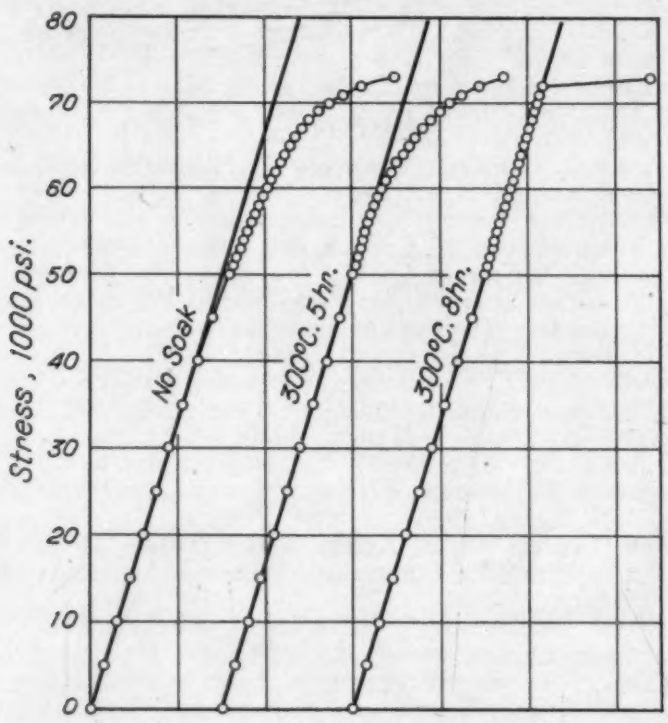

Fig. A—Stress-Strain Curves from Forged and Cold-Worked Cylinders.

is a definite difference between the two. The fact that a material exhibits considerable hysteresis with no evidence of proportionality of stress to strain is no indication that it is not still perfectly elastic in the sense that upon release of pressure applications very close to the final cold working pressure, the material will still return to its original after cold work dimensions. The real effect of the low temperature or stabilizing treatment is to bring about a return of proportionality of stress to strain. This is well illustrated in Figs. A and B. Fig. A shows typical stress-strain curves obtained from test specimens from a forged cold-worked cylinder in the cold-worked condition, and after stabilizing for 5 and 8 hours. The curve noted as unstabilized, shows the characteristic hysteresis looping with a very low proportional limit. Stabilizing for 5 hours at 300 degrees Cent. (570 degrees Fahr.) has released a part of the hysteresis with a resultant increase in proportional limit. The additional three hours or total of eight hours soaking has completely stabilized

the material and produced a clearly defined proportional limit which is practically equal to the elastic strength of the material. Further and more comprehensive evidence of the results of stabilization is indicated from the curves shown in Fig. B. These curves are plotted stress versus residual strain from readings obtained by the use of extremely sensitive instruments, and show quite clearly that stabilization after cold working merely relieves hysteresis and brings out a clearly defined proportional limit point. In this connection

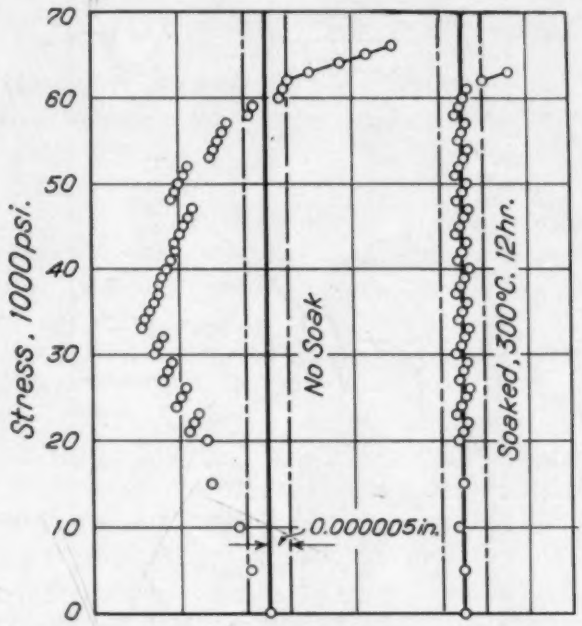

Fig. B—Results of Stabilization of Cylinders.

it may be well to note that the time and temperature required for complete stabilization will be found dependent to a large degree on the composition.

Regarding the remarks in the first paragraph, page 217, contrary to the author's statement, we have found that so long as the temperature used is just sufficient for stabilization and not high enough to produce a change in material microstructure, the only change in physical properties produced is the elimination of hysteresis looping with the accompanying development of a clearly defined proportional limit, no other property being affected.

In connection with the general subject of cold working it may be of interest to note that the increased elastic strength produced by cold working either a hollow cylinder by radial expansion, or a rod or bar by longitudinal stretching, can be readily predetermined from the true-stress curve obtained from the tension test of a single specimen of the material, as shown in Fig. C. These were obtained from a cylinder before cold work, and after cold work and stabilization. It will be noted that the curve of the after cold work and soak condition superimposes on the before cold work curve, at a distance to the right equivalent to an amount of stretching or cold working of the before cold work specimen equal to that given the cylinder at the position from which the specimen was taken. The new proportional limit point indicating the increased elastic strength produced. From such an initial curve it is

possible to determine the elastic strength at any part of a cylinder wall provided the percentage permanent enlargement at that particular point is known.

The curves noted in Fig. C were obtained from cast material which accounts for the large drop at the yield point. It has been our experience that this drop is definitely associated with the hot-worked condition of the material, that is, the presence of porosity, voids, etc. The condition when present is

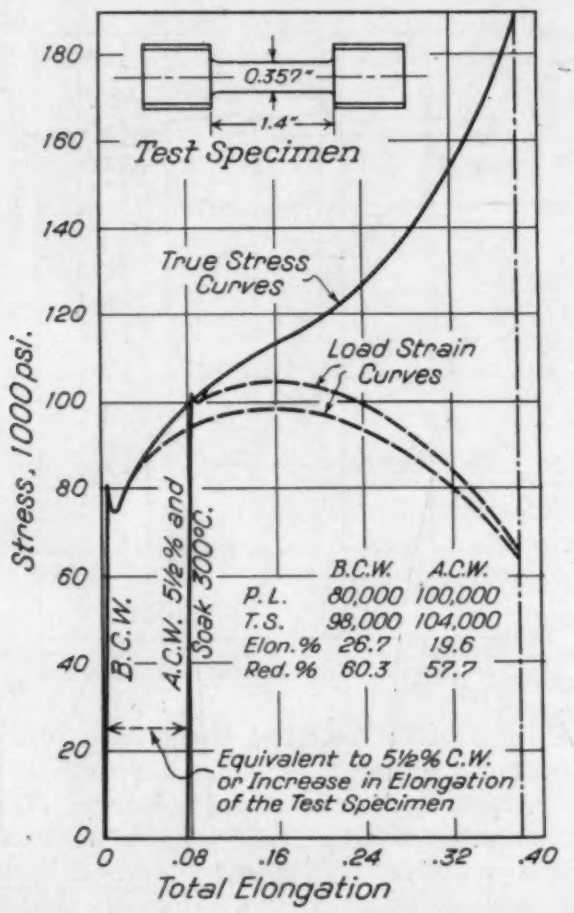

Fig. C—Curves from Cylinders of Cast Material Before and After Cold Work and Stabilization.

greatly reduced in magnitude and many times completely eliminated by slight cold working.

In conclusion it is desired to point to the fact that for all the large amount of investigative work which has been done on the subject of cold working hollow cylinders, the true reason for the increased elastic strength produced has not as yet been conclusively established. Further research into the new theories regarding the plastic deformation of crystals and the phenomena of stabilization by low temperature heat treatment, will undoubtedly lead to a clarification of the entire subject. The paper as presented, while it represents a review or survey of available literature, can only therefore be considered as a summation of ideas which may be subject to change at a later date.

INVESTIGATION OF FATIGUE STRENGTH OF AXLES PRESS-FITS, SURFACE ROLLING, AND EFFECT OF SIZE

By T. V. Buckwalter and O. J. Horger

Abstract

This paper presents fatigue test data on 2-inch diameter axles to show the great weakening effect on the axle due to the mounting of press-fitted members. It is shown that a press-fitted member reduces the fatigue strength of the axle from ½ to ⅓ of its potential strength available without the fitted member. Means of improving this weakening effect are described, whereby the fatigue strength of the usual press-fitted assembly may be made more than twice as strong by surface rolling the axle at the press-fit seat.

The effect of surface rolling on large size axles and how the fatigue strength of specimens, having different forms of stress concentration, vary with size is introduced with an interpretation of the present knowledge of "size effect" for discussion purposes. Announcement is also made of the construction by the Timken Roller Bearing Company of a large fatigue machine having sufficient capacity to test 13.5-inch diameter axles as a means of investigating "size effect," heat treating, forging, and problems associated with large sections.

IT WAS the purpose of this investigation (1) to determine the weakening effect caused by the presence of press-fitted members on the fatigue strength of axles and (2) to investigate means of strengthening axles at the press-fitted section, in particular to study the strengthening obtained by surface rolling the axle. This problem is of importance in many fields. Press-fit assemblies or shrunk-on members are of frequent occurrence, and due to increased severity in service conditions many failures have resulted in recent years. This is true for railroad or automobile axles at wheel or bearing seats, crank-pins of locomotives, crankshafts, propeller shafts and other ap-

A paper presented before the Eighteenth Annual Convention of the American Society of Metals, Cleveland, October 19 to 23, 1936. Of the authors, T. V. Buckwalter is vice president, and O. J. Horger is research engineer, Timken Roller Bearing Co., Canton, Ohio. Manuscript received July 23, 1936.

plications of press-fit assemblies. (3) It was also desired to determine the effect of size of fatigue test specimens.

Most fatigue tests are made with specimens that are small compared with the size of machine parts used in service. It is of interest to know how reliable the small specimens are for the determination of the strength of large specimens.

Press-Fit Tests

Three different types of fatigue tests, types A, B, and C, were conducted in order to investigate the fatigue strength of press-fit

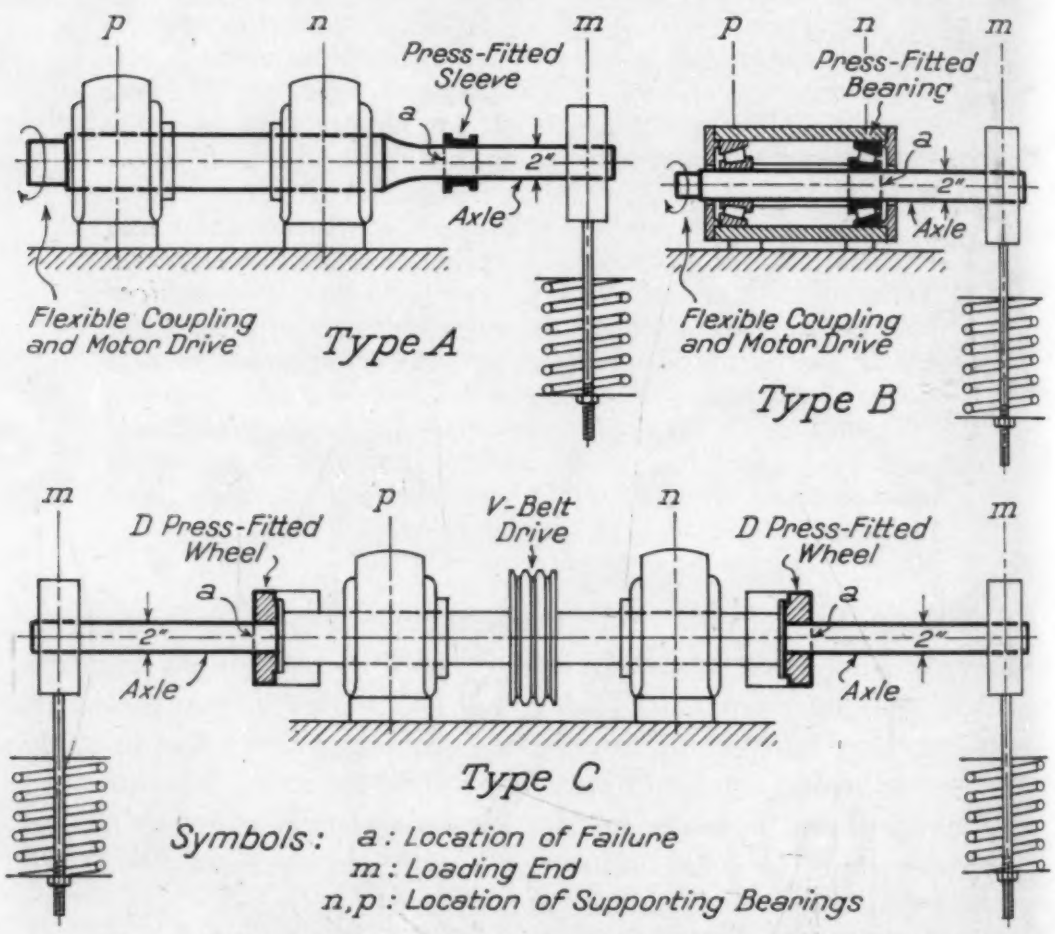

Fig. 1—Diagrams Showing the Three Types of Press-Fit Assemblies Tested.

assemblies under various load conditions. In all three cases the axles are 2 inches in diameter at the critical section and are tested in rotating bending machines. The line diagrams in Fig. 1 show the test conditions.

Type A simulates effect of ring pressed or shrunk on shaft.

Cantilever type machine with axle loaded at m and supported at sections n and p. A sleeve (inner race of Timken Roller Bearing without rollers) is press-fitted on axle and section "a" indicates general location of axle fatigue fracture.

The reduction in strength of axle is due to effect of press-fit alone, since no load is transmitted through the sleeve.

Type B simulates action of anti-friction bearing on shafts.

Cantilever type machine loaded at m and supported at n and p. At n the inner race of the roller bearing is press-fitted on axle and axle fatigue failure develops at "a."

The reduction in strength of axle is due both to press-fit effect and effect of load taken through the bearing. The effect of the press-fit alone was determined from the tests of Type A.

Type C simulates action of wheel press-fitted on axle and submitted to thrust and radial load at rim.

The axle is loaded at m and held in wheel D by press-fit. Wheel D is bolted to a main spindle supported at n and p. Axle fatigue failure occurs at "a."

The reduction in strength is due to effect of press-fit and to the action of vertical load and crushing moment which must be transmitted through the press-fitted connection.

The stresses mentioned in this paper are nominal bending stresses in axle at end face of sleeve or wheel where axle failure occurs, and are obtained from the beam theory:

$$\text{Bending Stresses } S = \frac{M\ c}{I}$$

M = Bending Moment (Load X lever arm)
I = Moment of Inertia of Axle.
c = Radius of Axle.

Axle Material Tested

Test results obtained with S.A.E. 1045 steel from the three types of tests mentioned above and results on 2.75 per cent nickel steel tested in machines of type B and C are given in the present paper. The S.A.E. 1045 steel was normalized and drawn while the nickel steel was quenched and tempered. The chemical composition and physical properties are given in Table I.[1]

From tests made on polished 2-inch diameter plain specimens

[1]The S.A.E. 1045 steel is the same material as used in a previous investigation reported in reference 1 of Bibliography.

Table I

Materials Used for Test Axle

	C	Mn	P	S	Si	Cr	Ni
S.A.E. 1045	0.45–0.49	0.67–0.77	0.013–0.017	0.031–0.037	0.20–0.26	0.03	0.04–0.06
2.75 per cent nickel	0.24	0.86	0.034	0.021	0.24	....	2.79

Heat Treatment—S.A.E. 1045—Normalized 1620 Degrees Fahr. and Drawn 1115 Degrees Fahr. 2.75 Per Cent Nickel—Steel Quenched 1475 Degrees Fahr. and Tempered 1150 Degrees Fahr.

Physical Properties

	S.A.E. 1045	2¾ Per Cent Ni
Yield Point, Pounds Per Square Inch	47,800	86,250
Ultimate Strength, Pounds Per Square Inch	88,800	111,000
Elongation in Per Cent	32	23
Reduction in Area in Per Cent	48.5	60
Brinell Hardness	177	240

it was found that the fatigue strength in bending of the S.A.E. 1045 is 34,400 pounds per square inch. (1)[2]

Material of Press-Fitted Sleeve and Wheel

The sleeves used in tests of type A and B were standard inner races of Timken roller bearings being 2-inch bore by $2\frac{1}{16}$ inches long.

The wheels used in type C tests were of forged steel in accordance with the recommendations for railroad wrought steel wheels by the Association of American Steel Manufacturers. The specifications for chemical analysis are as follows:

C	Mn	Si	S	P
0.65-0.85	0.60-0.85	0.15	0.05	0.05

The physical properties are:

Ultimate Tensile Strength .....110,000 pounds per square inch
Yield Point ..... 60,000 pounds per square inch
Brinell Hardness 250

Test Machines

Type A and B. Details of these machines have been described in previous papers (2), (3). Only results necessary for the present discussion will be mentioned here.

Type C. The test machine is shown in Fig. 2. It consists of

[2]The figures appearing in parentheses refer to bibliography given at the end of the paper.

a main spindle $3\frac{3}{16}$ inches in diameter supported by two pillow blocks equipped with Timken roller bearings. A hub is pressed on each end of the main spindle and to these hubs are bolted the wheels into which the test axles are press-fitted. Thus, on each machine, two axles may be tested at different bending stresses. Three of these machines were built.

The spindle is driven through a 5 horsepower motor and a V-belt drive. The pulley is located between the pillow blocks. Two sizes

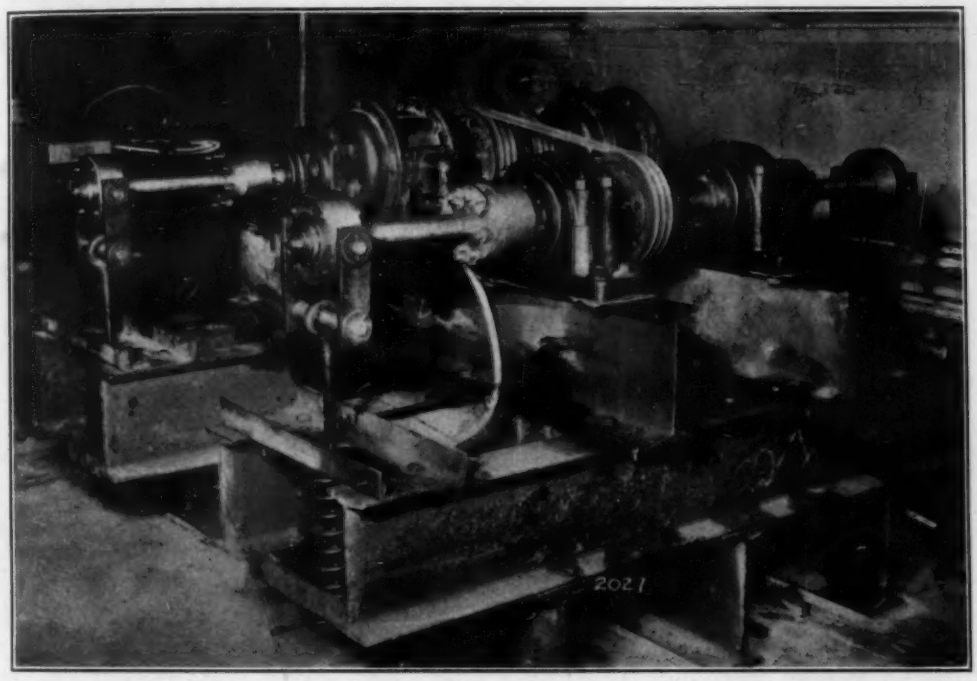

Fig. 2—Photograph of Fatigue Testing Machine. Type C.

of motor pulleys are provided which give a speed of 1750 revolutions per minute used for light loads on the test specimens and 1290 revolutions per minute for high loads.

The load is applied at the end of the test axle by calibrated springs (300 to 400 pounds per square inch per spring), acting on a pull rod and connected to a trunnion arrangement mounted on a Timken bearing on the end of the test axle.

The dimensions of the press-fit assembly are shown in Fig. 3. The press-fitted portion of the axle is smooth turned and given a light fine file finish. The amount of tight fit between axle and wheel assembly was determined with an accuracy of $\pm$ 0.0001 inch. The maximum pressure required to press-on the wheel was recorded, and

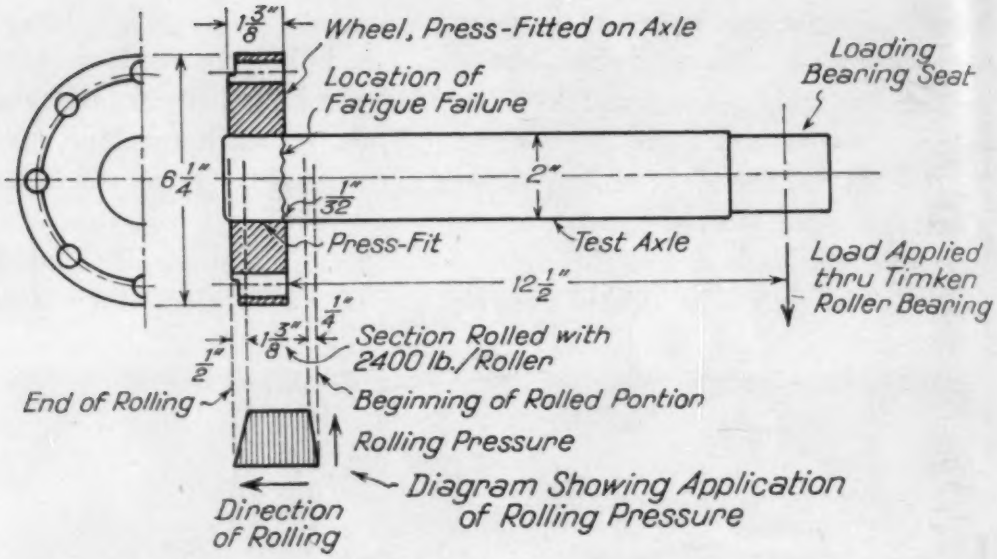

Fig. 3—Dimensions of Test Axle and Press-Fitted Wheel.

the assembly was then placed in the lathe and the wheel faced off so as to assure the running true of the axle when placed in the fatigue machine.

Test Results

(a) *Axles without Surface Rolling.*

The test results obtained in machines of type C are given in Table II and plotted in Fig. 4. A typical failure is shown in Fig. 5.

Table III summarizes the results obtained on machines A, B and C. It is seen that in spite of the differences in press-fit conditions, the strength of an axle is reduced to 35 to 44 per cent of its potential value available without the press fitted member.

Other investigators have made somewhat similar tests on smaller diameter specimens and found similar results (4, 5, 6).

Table II

Test Results Obtained in Machine Type C (Fig. 1) Axles S.A.E. 1045 Steel Not Surface Rolled

Spec. No.	Press Fit on Diameter Between Wheel and Axle	Nominal Bend. Stress Pounds Per Square Inch	No. of Cycles in Millions	Remarks
102	0.0017 inch	19,500	3.1	Failed
107	0.0017 inch	15,000	25.8	Failed
101	0.003 inch	13,500	127.5	Did not fail
118	0.004 inch	14,500	70.0	Did not fail

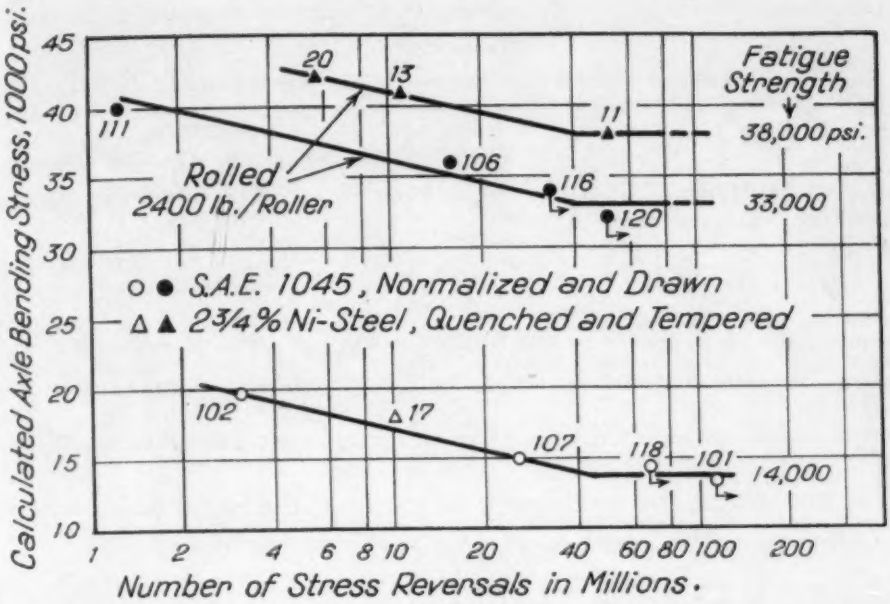

Fig. 4—Fatigue Test Data on Press-Fit Study of 2-Inch Diameter Axles.

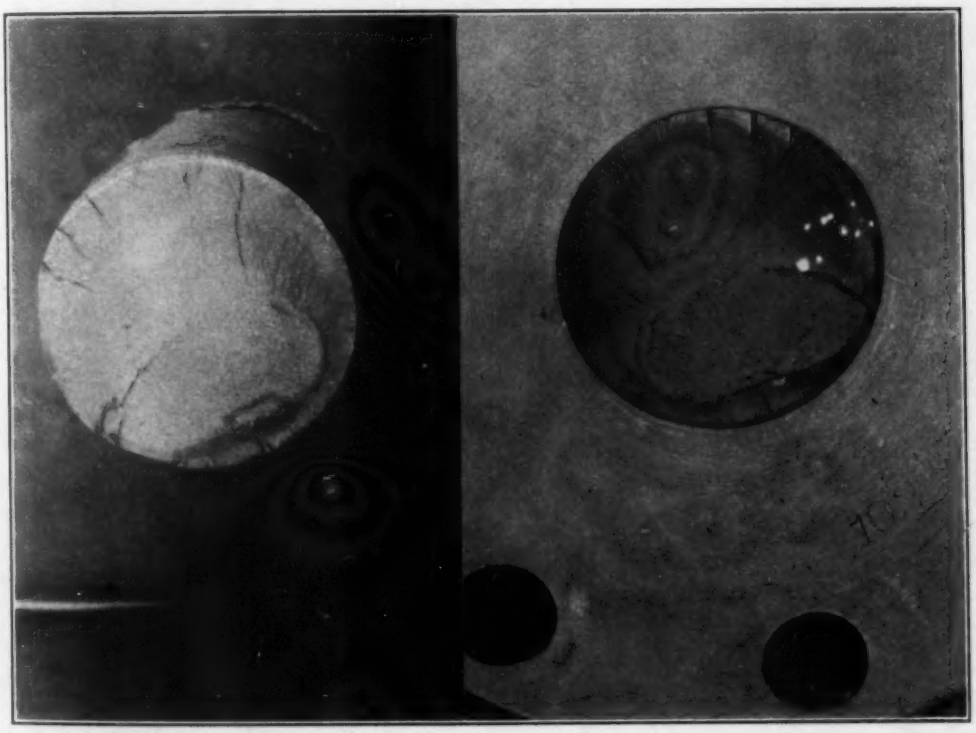

Fig. 5—Photograph of Typical Fatigue Failure.

The action of the press-fit conditions in type A test reduces the strength to 44 per cent. If a force or a moment are taken through the press-fit assembly as in test type B or C, the axle is still further weakened.

Table III

Fatigue Strength of Axles—Not Surface Rolled

Type of 2" Axle Test	S.A.E. 1045 Normalized and Drawn		2.75 Per Cent Nickel Steel Quenched and Tempered	
	Pounds Per Square Inch	Per Cent	Pounds Per Square Inch	Per Cent
Plain Specimen, without press fit	34,000[3]	100	48,000[4]	100
Type A, Fig. 1	15,000[3]	44	.....	...
Type B, Fig. 1	12,000[3]	35	Less than 17,000	Less than 35
Type C, Fig. 1	14,000	41	Less than 17,000	Less than 35

[3]Data previously published (2, 3).
[4]Estimated for 2-inch specimen—The fatigue strength in bending of this steel is 57,000 pounds per square inch for 0.3-inch specimen. Due to the size effect discussed at the end of this paper, the fatigue strength is reduced for 2-inch specimens. A reduction of 15 per cent is assumed here.

It is noted that the weakening action is of the same order whether carbon steel or nickel steel is used as material for the axle.

Surface-Rolled Axles

In the surface rolling operation, rollers are pressed against the surface of the axle (Fig. 6) so as to plastically deform the surface layers of the axle. The axle which is placed between lathe centers is rotated and the rollers that are mounted on the lathe carriage are slowly moved along the length of the specimen.

This process is similar to the burnishing operation that is used in railroad shops to obtain a wear resisting surface for plain bearing journals. The importance of surface rolling for increasing the fatigue strength of axles was first recognized in Germany in 1928 by O. Föppl (7).[5] In this country it has also been investigated in particular by the Westinghouse Research Laboratories and by the Timken Roller Bearing Company.

In the tests presented here four rollers as shown in Fig. 6 were used. Fig. 7 is a picture of the rolling device in operation.

Preparation of Surface-Rolled Axle

The surface rolled axles do not differ in design from the axles previously described except that the portion of the axle press-fitted into the fitted member is surface-rolled. Before rolling, the axle is smooth turned and given a fine file finish to the diameter that will give the proper fit. The reduction in diameter of 2-inch axle due to

[5]An historical account of surface rolling was given in a previous paper (1).

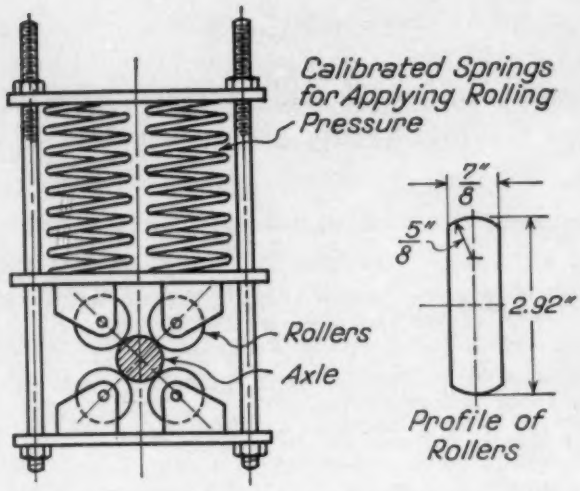

Fig. 6—Diagrammatic Arrangement of Device for Surface Rolling.

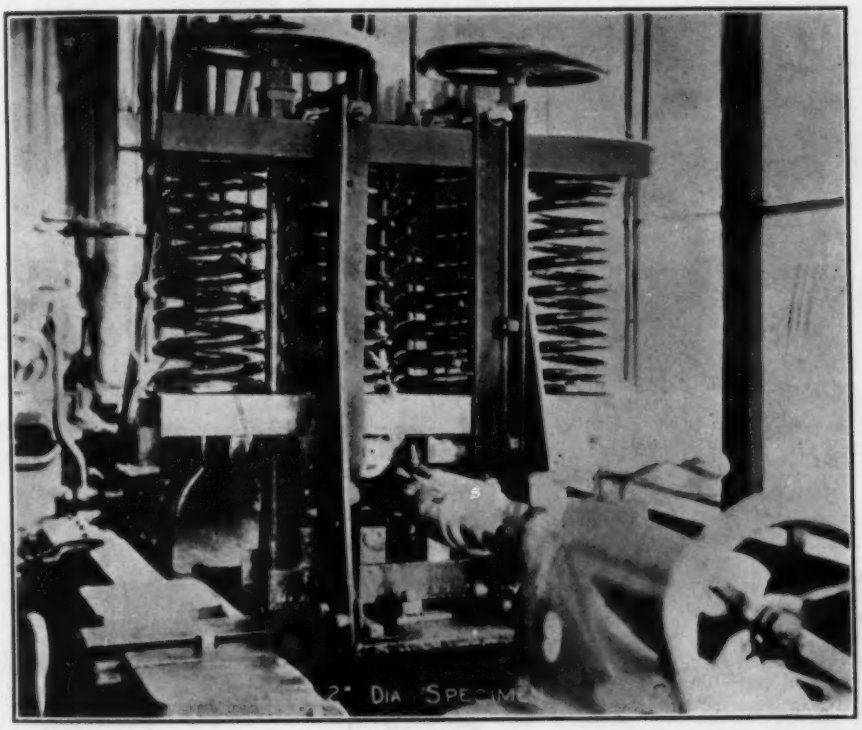

Fig. 7—Photograph of Surface Rolling Device Set Up in Lathe.

rolling is about 0.0004 inches per 1000 pounds load per roller for a fine file finish previous to rolling. No further machining of the rolled portion is done after the rolling operation. Axles have also been ground previous to rolling and in such cases the better surface conditions result in a smaller reduction in diameter due to rolling.

Test Results

The test results obtained with surface-rolled axles tested in machines of type C (Fig. 1) are given in Table IV and plotted in Fig. 4.

Table IV

Test Results Obtained in Machine Type C (Fig. 1) Press-Fitted Portion of Axle Rolled at 2400 Lbs. Pressure Per Roller

Spec. No.	Press-Fit on Diameter Between Wheel and Axle	Nominal Bend. Stress Pounds Per Square Inch	Number of Cycles	Remarks
	S.A.E. 1045—Normalized and Drawn			
111	0.004 inch	40,000	1,250,000	Failed
106	0.004 inch	36,000	15,600,000	Failed
116	0.006 inch	34,000	33,400,000	Did not Fail
120	0.0055 inch	32,000	51,800,000	Did not Fail
	2.75 Per Cent Nickel-Quenched and Tempered			
20	0.003 inch	42,000	5,600,000	Failed
13	0.003 inch	41,000	10,900,000	Failed
11	0.005 inch	38,000	51,800,000	Did not Fail

The data for the surface rolling is:

Roller as shown in Fig. 6.

Roller pressure: 2400 lbs. per roller.

Feed: 63.8 Threads per in.

R.P.M. Lathe spindle: 22.

The increase in fatigue strength due to surface rolling is given in Table V.

Table V

Comparison of Fatigue Strength Unrolled and Rolled Axles Type C (Fig. 1)

Type of 2-Inch Axle Test	S.A.E. 1045 Normalized and Drawn Pounds Per Square Inch	Per Cent	2.75 Per Cent Nickel Quenched and Tempered Pounds Per Square Inch	Per Cent
Not Surface Rolled	14,000	100	17,000	100
Surface Rolled 2400 lbs. per roller	33,000	236	38,000	225

Some difficulties arose in these tests. The first test was made with a press-fit of 0.0017-inch and a bending stress of 32,000 pounds per square inch. As soon as the specimen was set in rotation the axle pulled out of the wheel. In the next test the fit was increased to 0.003-inch but after running for about 12 million revolutions the axle worked itself loose in the wheel. A further increase in press-fit was useless because the press-fit stresses together with the crushing

stresses caused by the moment transmitted from the axle into the wheel result in yielding of the material around the bore of the wheel. This yielding is sufficient to produce loosening of the shaft and to prevent its runing for any great length of time.

When testing the conventional press-fit assembly, where the axle is not rolled, it was found that the axle was weaker than the wheel and fatigue failure developed in the axle. When the axle is surface-rolled, however, the axle becomes stronger than the wheel and in order to develop axle failures in test, to determine the increased axle fatigue strength due to surface rolling, it becomes necessary to strengthen the wheel. The wheel may be strengthened by:

(a) Increase yield point strength of the wheel material around the bore. This may be done by plastically deforming this material before the axle is pressed on by pushing an over-sized tapered plug through the wheel bore.

(b) or to increase the length of the wheel so as to reduce the crushing stresses caused by the bending moment which is transmitted from the axle into the wheel.

The first method was successfully used in the tests reported here.

Conclusions

From tests made on the various types of press-fit assemblies shown in Fig. 1, it may be concluded:

(1) a press-fit member reduces the fatigue strength of an axle to less than half of the strength of a similar axle not having a press fitted member.

(2) By surface rolling the axle at the press-fitted section, it is possible to more than double the fatigue strength of the usual press-fitted axle assembly. Thus surface rolling permits the use of press-fitted assemblies having about the same strength available in plain axles which do not have a press-fitted member.

Size Effect

The tests reported here are on 2-inch diameter axles, and the question is now: (1) will the favorable effect of surface rolling be as great for larger size axles and (2) how does the fatigue strength of specimens vary with size of specimen having different forms of stress concentrations. In order to answer these questions and others relating to forging and heat treating practice and the general prob-

lems associated with large sections the Timken Roller Bearing Company is building a fatigue testing machine with a capacity sufficient to test axles of 13½-inch diameter.[6]

The question of size effect is of importance, but it is a phenomenon that has been little studied and on which few test data exist. Here an interpretation of the present knowledge on this subject is presented for discussion.

The test results may be generalized by stating that the fatigue strength of the material depends upon the stress distribution or stress gradient in the region of maximum stress. If the stress decreases rapidly from its maximum value the material does not fail as readily as when the same maximum stress values vary little over the cross section of the machine part.

A relation between maximum allowable stress and stress gradient is determined here from fatigue test data on shafts with fillets (Fig. 8a). It is not necessary to know the absolute value of the stress gradient in order to obtain a comparison of test data but it is sufficient to assume that one specimen with a certain diameter has a gradient "unity." The gradient for other specimens can be obtained by comparing their stress distribution with the distribution of the one with gradient unity.

The curves shown in Fig. 9 give this relation for two different types of carbon steel. The stress gradients were obtained by making a photoelastic study of the stress distribution at the point of stress concentration of the fillet. The fatigue data are those published by Lehr and Mailander (8) in Germany and Peterson and Wahl (9) in this country. The value of stresses plotted in Fig. 9 are maximum stresses at the fillet, that is nominal stress multiplied by stress concentration factor. On the stress gradient axis, the ratio of stress gradient to the unity stress gradient is plotted. The shaft with unity gradient is:

$$d = 1.18'', \; R/d = .5, \; D/d = 2$$

The ratios of stress gradients are obtained by photoelasticity. In Fig. 10, a fringe photograph of the tension side of a beam in pure bending is shown for the case $R/d = .062$ and $D/d = 2$. The stress distribution toward the inside of the shaft and at the point of maximum stress are shown in Fig. 8b for three different fillet ratios ($R/d = 0.062$, 0.104 and 0.5).

[6]This machine is of the type C described in this paper (Fig. 1).

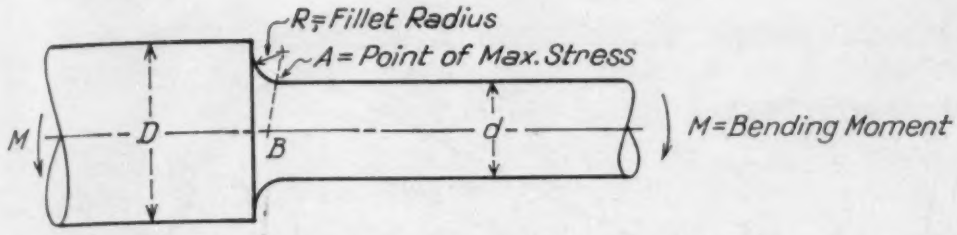

Fig. 8a

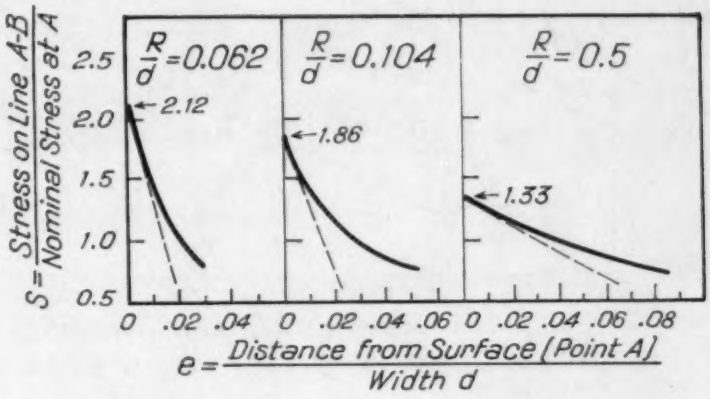

Fig. 8b

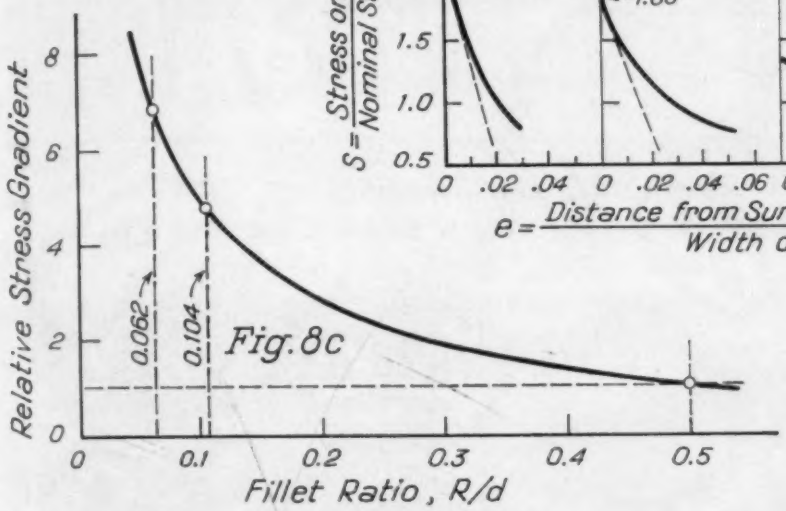

Fig. 8a—Axle With Fillet Submitted to Bending.
Fig. 8b—Stress Distribution Along Line A-B Obtained by Photoelasticity.
Fig. 8c—Curve of Relative Stress Gradient—Fillet R/d = 0.5 Has Stress Gradient "Unity."

The gradient of these curves are plotted in Fig. 8c. From the curve connecting these points it is possible to obtain the gradient of any other fillet ratio.

In order to correlate the photoelastic data with fatigue tests, it is to be noted that in a shaft the stress does not vary in one direction only. This fact has to be considered when the stress gradient of any particular shaft is compared to the stress gradient unity.

When comparing shafts of same diameters, we assume:

(a) Normal to the surface the stress gradient varies as shown in Fig. 8c for different fillet ratios (b) In the axial direction along the surface of the shafts it is assumed to vary in the same manner (c) Circumferentially the stress gradient is assumed to be the same for all fillet ratios.

If the shafts are of different diameters but with the same R/d

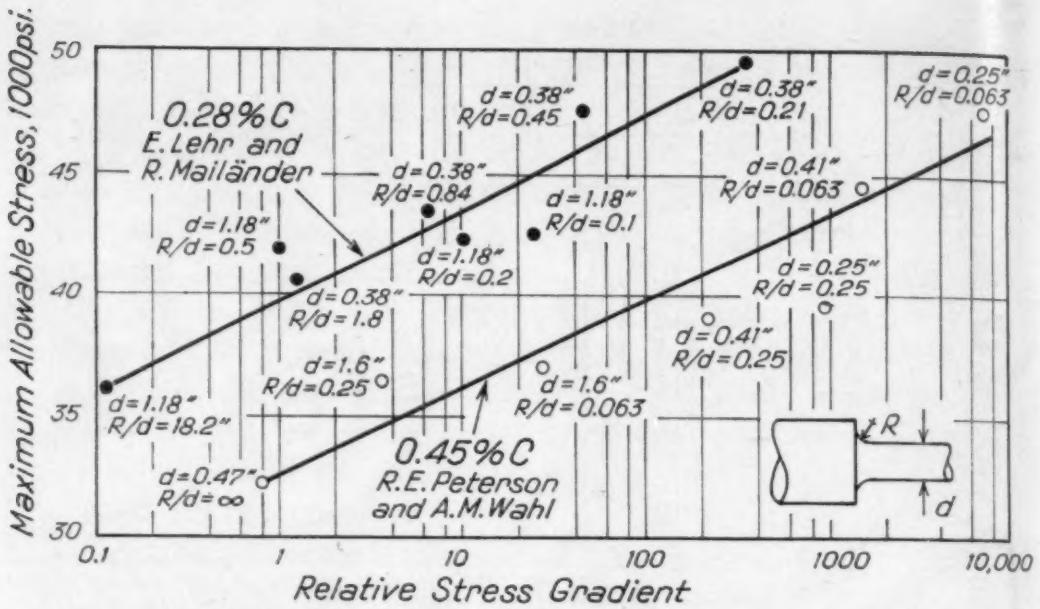

Fig. 9—Relation Between Maximum Allowable Bending Stress and Stress Gradient. —Shafts with Fillet.

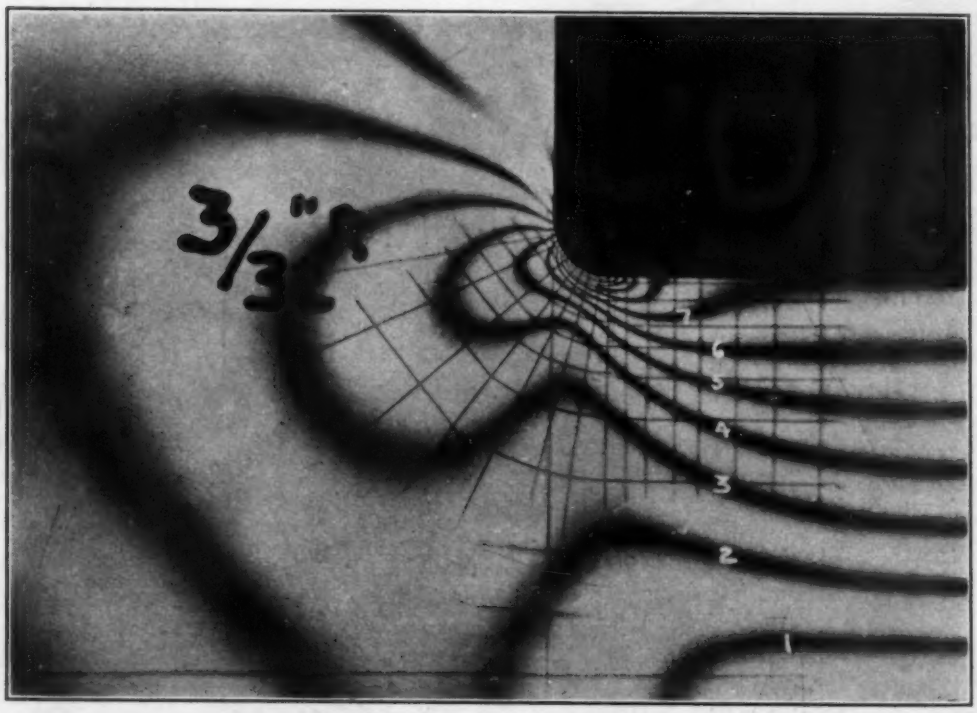

Fig. 10—Fringe Photograph for R/d = 0.062 and D/d = 2.

then the stress gradient is assumed to vary as the cube of the ratios of the diameters.

From the curves of Fig. 9 it is possible to determine the maximum stress that a machine part may safely withstand in fatigue

when the stress distribution is known and the stress gradient may be determined. In this manner the "sensitivity factor" that was introduced a few years ago (10, 11) is not used. The sensitivity factor had been introduced with the thought that it would be a constant factor of one material. Recently R. E. Peterson and A. M. Wahl (9) have shown that this is not the case.

These curves are presented here only as a suggestion for further research in that direction. Further test data are required especially with alloy steels in both small and large sections and carbon steels in large sections before more definite conclusions may be reached.

Acknowledgments

We acknowledge the beneficial discussions had with Dr. S. Timoshenko and to Professor E. L. Ericksen for the accommodations provided in making these tests at the University of Michigan. The assistance of Dr. J. L. Maulbetsch, formerly with The Timken Roller Bearing Company, is also acknowledged.

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DISCUSSION

Written Discussion: By T. S. Fuller, Research Laboratories, General Electric Co., Schenectady, N. Y.

This paper by Messrs. Buckwalter and Horger constitutes another valuable contribution on the subject of fatigue, and will now be added to the long and excellent list of papers on this subject already published by these authors. The importance of a study of the fatigue strength of axles, press-fits, surface rolling and effect of size can hardly be overestimated.

The shape of the stress cycle graphs resulting from work which has been done recently on the fatigue strength of rotating members, in which localized stresses, due to one thing or another, have been present, strongly suggests that under such conditions there may be no definite fatigue limit.

The lower curve of Fig. 4 of the present paper, showing the stress cycle graph of S.A.E. 1045 steel in the normalized and drawn condition and with a press fitted collar is gratifying because the shape suggests the existence of an endurance limit under conditions of localized stress and the possibility of its determination on the basis of one hundred million cycles. At any rate it is certain that the knowledge concerning the subject in question can be most rapidly advanced by the accumulation of data from tests which have been run to one hundred million cycles or more.

Those close to the subject will await with interest the results obtained by the large fatigue machine, announced in the paper, and having sufficient capacity to test axles 13½ inches in diameter.

Authors' Reply

Mr. Fuller raises an important question as to whether a definite endurance limit may be established. Our fatigue tests are being extended to 200 to 300 million cycles in order to determine if the fatigue strength changes much over the value at 100 million cycles.

Present indications are that the change will be small, but it is yet too early to come to definite conclusions.

The S-N curves in the paper are, in most cases, based on the cycles required to actual fracture of the 2-inch axles, but of more practical interest is the cycles to the development of the first crack. This is for the reason that railroad axles are regularly inspected and scrapped at the first indication of a crack. At bending stresses, near the endurance limit of unrolled 2-inch test axles, it has been found that a fatigue crack may develop but progresses very slowly. In this connection it should be mentioned that Moore[1] has reported a fatigue crack in axle steel specimens would not progress when stressed to 50 per cent of the endurance limit of the virgin steel. In the case of the press-fitted members reported in this paper, it is being suggested here that the initial crack may develop due to rubbing corrosion and stress between the axle and wheel. The bending stress may then be too low to cause this crack to propagate. This phase of crack initiation and propagation is being actively studied.

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[1]H. F. Moore, "A Study of Fatigue Cracks in Car Axles," University of Illinois Bulletin 165, June 1927.

ENDURANCE OF GEAR STEELS AT 250 DEGREES FAHRENHEIT

By A. L. Boegehold

Abstract

Fatigue tests on notched rotating beam specimens flooded with oil at 250 degrees Fahr. showed that carburized and hardened gear steels were less resistant to fatigue than at room temperature. It is shown that under these test conditions a martensitic surface structure is more fatigue resistant than one predominantly austenitic. The decrease of endurance limit by application of hot oil is about 10 per cent. The fatigue life above the endurance limit in contact with hot oil is only about 10 per cent of that at room temperature.

IN A PAPER presented last year,[1] fatigue tests of rear axle gears in a rear axle assembly and fatigue tests made on notched rotating beam specimens were reported. In that paper, the relative order of durability for the different steels, as indicated by the notched bar fatigue test, did not agree satisfactorily with results of rear axle tests. It was presumed that this lack of agreement was because of variations in stresses depending upon how tooth bearing contacts were altered by distortion of gear teeth in heat treating. This explanation was not completely satisfying, however, and there remained a suspicion in our minds that some other unknown influence existed.

The question as to whether operating temperature affected the fatigue life had been considered many times but had always been dismissed as insufficient to account for very much change in properties other than possibly a slight strengthening. The temperature at which the surface of gear teeth operate has been estimated at about 250 degrees Fahr. judging from determinations which show temperatures around 180 degrees Fahr. in rear axle lubricating oil. It was at first reasoned that since the temperature of the rear axle gears in pas-

[1] "Rear Axle Gears—Factors Which Influence Their Life" by J. O. Almen and A. L. Boegehold. *Proceedings*, American Society for Testing Materials, Vol. 35, Part II, 1935.

A paper presented before the Eighteenth Annual Convention of the Society held in Cleveland, October 19 to 23, 1936. The author, A. L. Boegehold, is metallurgist, Research Laboratories Section, General Motors Corp., Detroit. Manuscript received June 23, 1936.

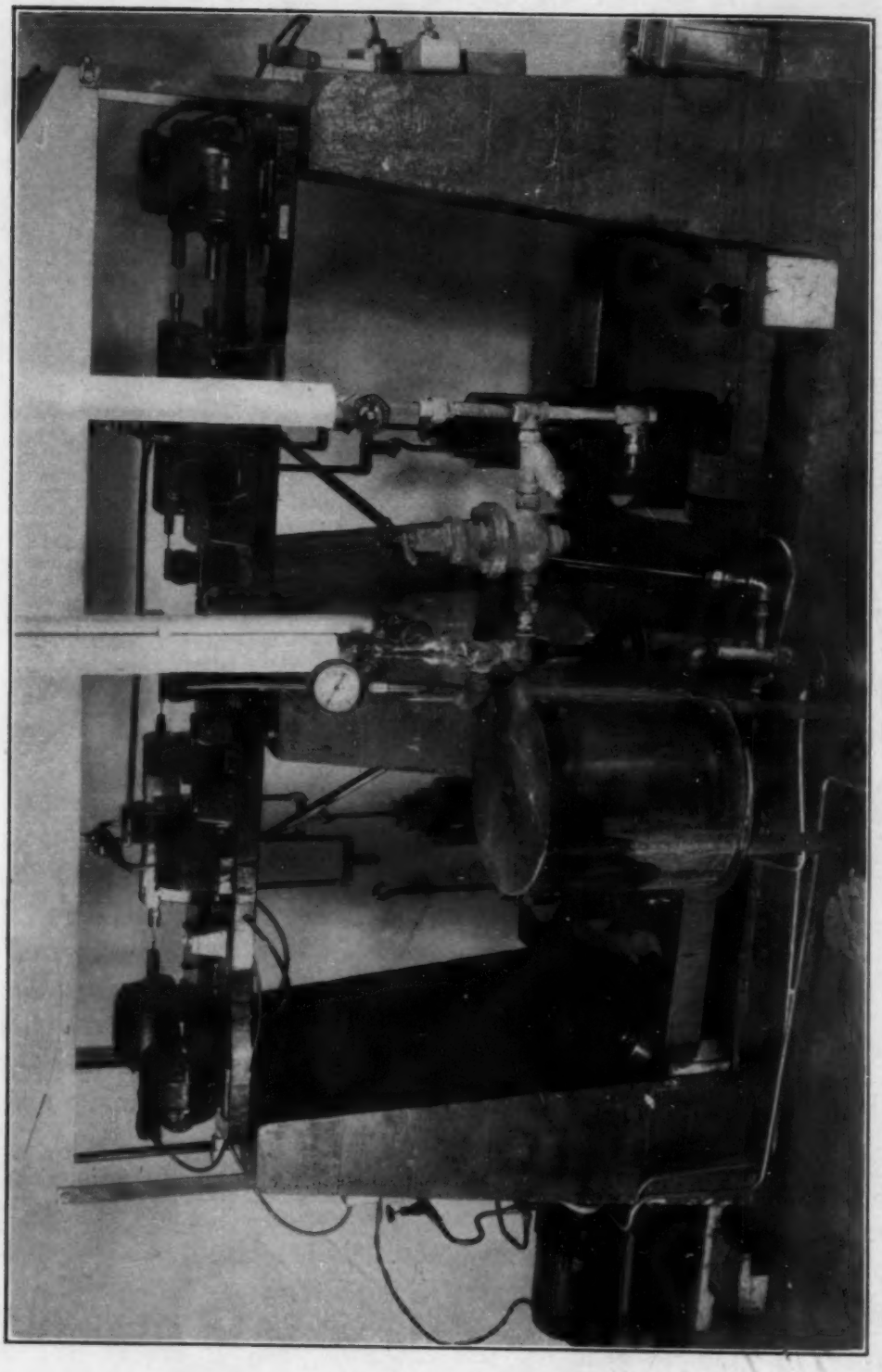


Fig. 1—Photograph of Apparatus Used in This Investigation.

senger cars are not high enough to affect their hardness, their strength would not be affected. Having observed, however, the effect of lubricating oils upon certain metals used for bearings, it began to

be apparent that it was not beyond the realm of possibility for highly stressed steel to be affected adversely by temperature and hot oil. It was decided, therefore, to investigate this possibility. Arrangements were made for conducting fatigue tests on notched rotating beam specimens flooded with lubricating oil at a temperature of 250 degrees Fahr. heated by means of a steam coil carrying steam at about 30 pounds per square inch pressure. Oil was pumped to the specimen and delivered at the center of the specimen at a temperature of 250 degrees Fahr. The apparatus is shown in Fig. 1.

Superla Bright Stock, which is the base sometimes used for blending rear axle lubricants, was used. This oil has a viscosity of 3450 seconds at 100 degrees Fahr. and 90 seconds at 250 degrees Fahr.

The specimen used for these tests is shown in Fig. 2. It will be noted that the notch in this specimen is a 100-degree notch as

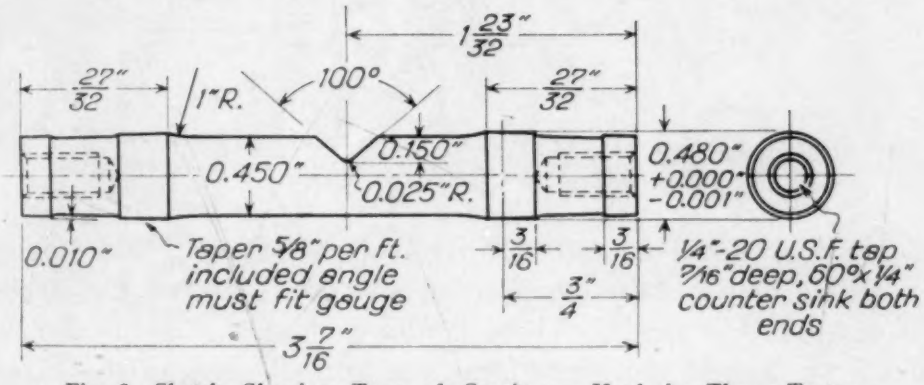


Fig. 2—Sketch Showing Type of Specimens Used in These Tests.

compared to 30 degrees used for test specimens discussed in paper previously cited. The 100-degree notch was adopted because it permitted a more uniform carburization at the bottom of the notch than could be obtained with the 30-degree notch. The amount of stress

Table I
Fatigue Limits for S.A.E. 3115 and 4615 Steels with Two Types of Notch

Steel	Treatment	Notch Angle	Endurance Limit Lbs./Sq. In.	Cycles Till Failure at 70,000 Lbs./Sq. In. Calculated Stress
4620	Oil quenched from	30	49,000	100,000
4615	carburizing box.	100	51,000	150,000
3115	300 Degrees Fahr.	30	40,000	50,000
3115	Draw.	100	46,000	100,000

concentration with the 100-degree notch was only slightly less than with the 30-degree notch as shown by data in Table I comparing

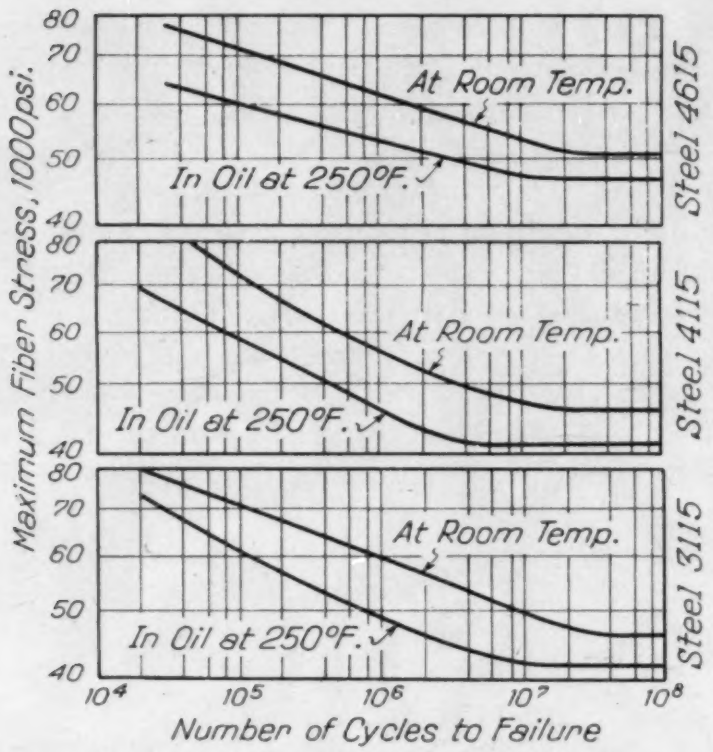


Fig. 3—Three Pairs of Curves Showing Effect on Endurance of S.A.E. 3115, 4115 and 4615 Steels Heated by Contact with Oil at 250 Degrees Fahr.

fatigue limits for 3115 and 4615 steels with the two types of notch.

Table II gives the composition and grain size of the steels on which tests were made. There is also a column showing the depth of case on the test specimens measured at the bottom of the V notch. Care was exercised to obtain as nearly as possible the same case depth on all test specimens. Carburizing and heat treatment procedure is also given in Table II.

Playing oil at 250 degrees Fahr. on the test specimen resulted in a decidedly lower fatigue resistance than was possessed by the steel at room temperature either dry or in contact with oil.

The three pairs of curves in Fig. 3 show how much the endurance of S.A.E. 3115, 4115 and 4615 steels was reduced by being heated by contact with oil at 250 degrees Fahr. The decrease in endurance limit is only 10 to 12 per cent but for stresses above the endurance limit, the fatigue life at 250 degrees Fahr. is only about 10 per cent of that at room temperature. An influence of this magnitude could easily be responsible for a large part of the disagreement between fatigue tests conducted at room temperature and rear axle gear fatigue tests.

Table II

Composition, Grain Size, Heat Treatment and Case Depth of Steels Tested

Steel	Analysis C	Mn	P	S	Ni	Cr	Mo	Grain Size	Heat Treatment	Case Depth Inches
4615	0.15	0.68	under 0.025	0.017	1.83	...	0.27	6–8	Carburized 9 hours at 1675° F. Quenched from carburizing heat in oil at 90-100° F. and drawn ½ hour at 300° F.	0.050–0.054
4115	0.15	0.51	under 0.025	0.006	...	0.82	0.22	6	Carburized 8 hours at 1675° F. Quenched from carburizing heat in oil at 90-100° F. and drawn at 300° F. for ½ hour.	0.050–0.051
3115	0.14	0.50	under 0.025	0.029	1.08	0.68	...	6	Same as 4115.	0.050–0.053
2515	0.16	0.65	0.035	0.017	4.92	...	...	7–8	Carburized 9 hours at 1675° F. Treatment (A) (For retaining austenite in case.) Quenched from carburizing heat in oil at 90-100° F. and drawn ½ hour at 300° F.	0.054–0.056
									Treatment (B) (For obtaining martensitic case.) Quenched from carburizing heat in oil at 90-100° F. Reheated to 1425° F. Quenched in oil at 90-100° F. and drawn ½ hour at 300° F.	
3312	0.12	0.40	0.014	0.016	3.91	1.35	...	6–8	Carburized 8 hours at 1675° F. Treatment (A) (For retaining austenite in case.) Quenched from carburizing heat in oil at 90-100° F. and drawn ½ hour at 300° F.	0.050–0.053
									Treatment (B) (For obtaining martensitic case.) Cooled in box. Reheated to 1500° F. Quenched in oil at 90-100° F. and drawn at 300° F. for ½ hour.	

In Fig. 4 are given the individual test data used for establishing curves in Fig. 3.

The question which immediately arises is what causes the decrease of endurance. Is it due to a softening from long exposure to

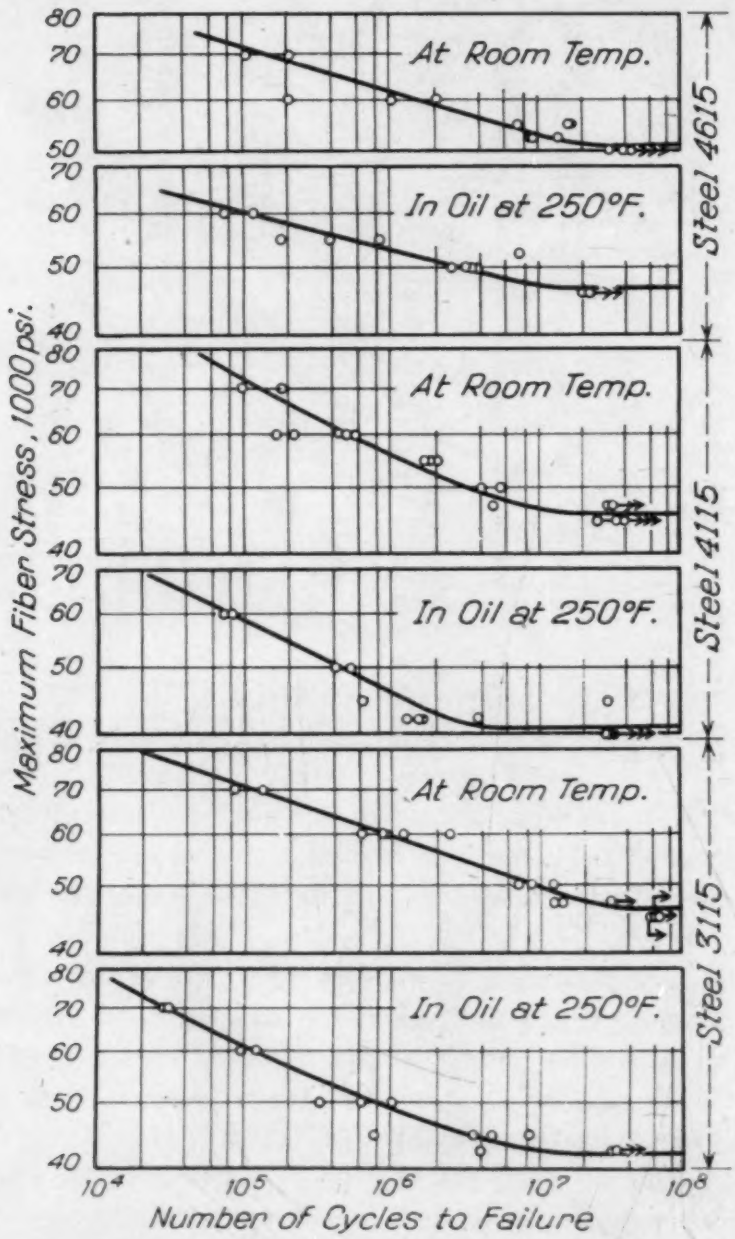


Fig. 4—Individual Test Data Used in Establishing Curves in Fig. 3.

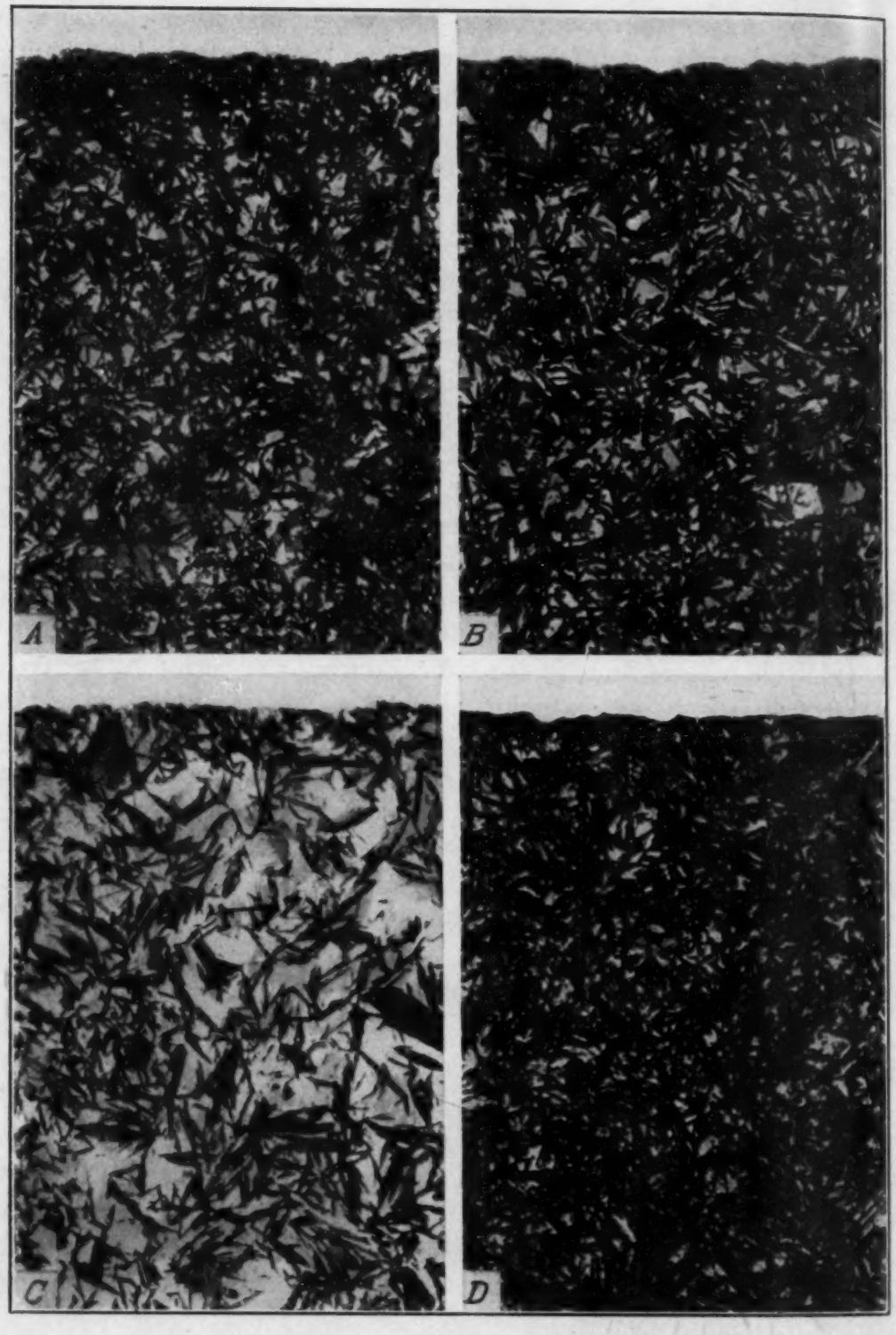
250 degrees Fahr. heat or is it the result of corrosion by organic acids in the oil formed as a result of oxidation of the oil at the elevated temperature?

Hardness tests made after heat treatment and after exposure to oil at 250 degrees Fahr. for the duration of the test showed no change of hardness. This is as would be expected because the specimens were all drawn at 300 degrees Fahr. before the test was made. File tests on the material before and after testing also showed no difference. Examination of the microstructure at the surface at a magnification of 500 diameters showed no change. Fig. 5, photomicrographs A and B, showing the microstructure of the box-quenched S.A.E. 4615 steel before and after the fatigue tests, are typical of all the steels tested and show no change due to long exposure to oil at 250 degrees Fahr. The structure shown in B is that of a specimen which had been run over 20 million cycles at 250 degrees Fahr. or for about 200 hours. That the decrease in fatigue life is not the result of a slow structural change is indicated by the fact that specimens that fail in about 15 minutes also show a proportional loss of fatigue life.

Examination of the surface of the test specimens under the microscope revealed no suggestion of corrosion or other change of appearance after exposure to the hot oil during the period of the test, consequently it does not seem to be a case of corrosion fatigue of the kind frequently reported in the literature. Thus, no evidence which would explain the loss of fatigue strength was obtained. There is a possibility that the loss of fatigue resistance is due to hydrogen embrittlement arising from the action of organic acids generated upon oxidation of the hot oil. The acid equivalent of the oil after being in the fatigue test apparatus for some time, however, was only 0.27 milligrams KOH per gram of oil. Whether serious hydrogen effect would arise from this degree of acidity is questionable.

It will be noted that the structure at the surface of the steels which were quenched from the carburizing box is preponderantly austenitic. To determine whether a preponderance of austenite or martensite at the surface of the carburized case influenced fatigue resistance of the steel in hot oil, S.A.E. 2515 steel was tested first with a preponderance of austenite at the surface obtained by box quenching after carburizing and then with a more martensitic surface layer obtained by quenching from the box and then requenching from 1450 degrees Fahr.

S.A.E. 2515 steel was selected for this test because the high nickel content aids in the formation of austenite at the surface when quenched from the box. It was found that the martensitic condition



Figs. 5A and 5B—Showing Microstructure of Box-Quenched S.A.E. 4615 Steel Before and After Fatigue Tests. × 500.

Figs. 5C and 5D—Show the Difference Between Structures of Carburized S.A.E. 2515 Steel Obtained by Box-Quenching and Subsequent Quenching from 1425 Degrees Fahr. × 500.

as the result of the double treatment was decidedly more fatigue resistant than the austenitic condition.

S.A.E. 3312 steel was tested in the same way and this steel also showed the martensitic condition to be more fatigue resistant when heated by contact with oil at 250 degrees Fahr.

Photomicrographs C and D of Fig. 5 show the difference between structures of carburized S.A.E. 2515 steel obtained by quenching from the box and by quenching from the box followed by a

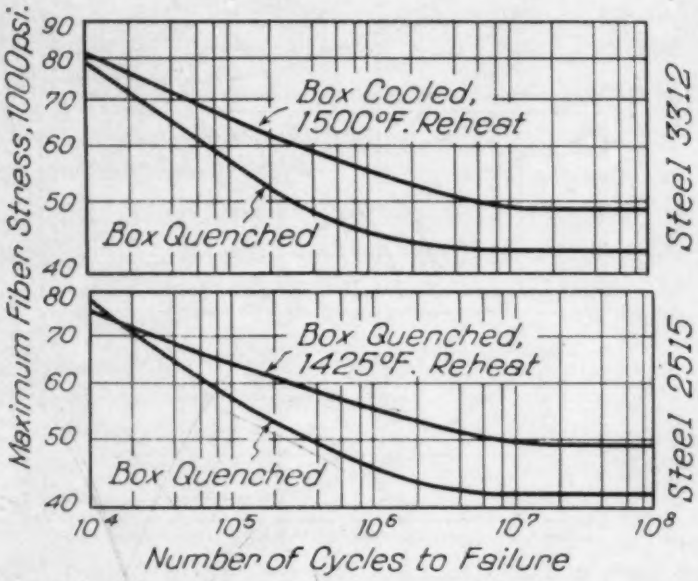


Fig. 6—Showing How Fatigue Resistance at 250 Degrees Fahr. in Contact With Hot Oil is Influenced by Heat Treatment of S.A.E. 2515 Steel and S.A.E. 3312 Steel.

quench from 1425 degrees Fahr. The two pairs of curves in Fig. 6 show how fatigue resistance at 250 degrees Fahr. in contact with hot oil is influenced by heat treatment. Curves in Fig. 7 show individual test data obtained in establishing these curves.

It will be noted in Fig. 6 that the curve representing the steel in the austenitic condition crosses the other curve in the high stress region indicating probably that work hardening of the austenite has made it more fatigue resistant than the martensite. Possibly a curve for an austenitic structure work-hardened before testing would be above the martensitic curve for all stresses.

These tests indicate that in evaluating steels for use in rear axle gears less chance of error in selecting the best steel will arise if selection is made on the basis of tests made at 250 degrees Fahr. in hot oil rather than on the basis of tests made at room temperature. The

life obtained by testing a steel at 250 degrees Fahr. should be a more accurate indication of the life to be expected from that steel in a rear axle gear. The number of cycles till failure in tests at room temperature are 700 to 1500 per cent more than when the steel is heated with oil at 250 degrees Fahr.

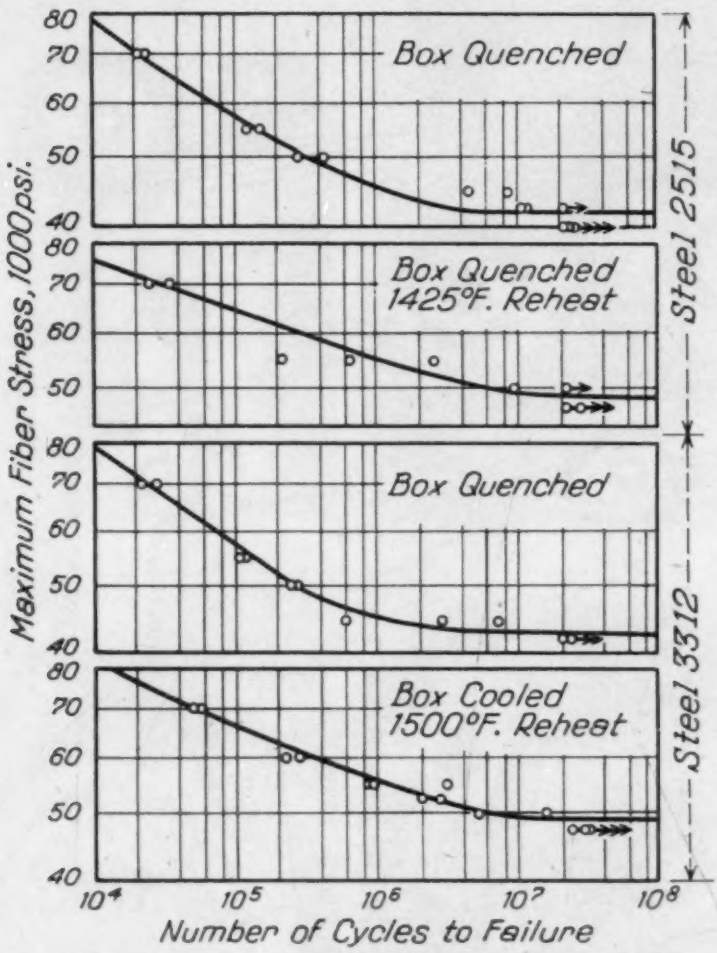


Fig. 7—Individual Test Data Used in Establishing Curves in Fig. 6.

Summary

Fatigue tests made on notched test bars heated by oil at 250 degrees Fahr. have resulted in the discovery that—

1. Resistance to fatigue of hardened carburized steel is considerably less at 250 degrees Fahr. than at room temperature.
2. The fatigue life in terms of cycles till failure, which is the usual way of measuring the life of rear axle gears, is as much as 1500 per cent greater at room temperature than at 250 degrees Fahr.

3. The structure obtained by box quenching after carburizing is considerably less fatigue resistant than that obtained by a double treatment including box quenching or cooling in the box followed by reheating to a temperature above the Ac_3 and quenching.
4. Work hardening of an austenitic structure appears to make it more resistant to fatigue in contact with hot oil.

The reason for the loss of durability at a temperature of 250 degrees Fahr. was not discovered, but the indications are that it is not a consequence of corrosion or of any structural change of sufficient magnitude to affect the indentation or file hardness nor the structure as revealed by a magnification at 500 diameters. The effect of hydrogen embrittlement, due to presence of organic acids, is mentioned as a possible explanation but supporting which there is as yet little evidence.

DISCUSSION

Written Discussion: By Arthur E. Focke, research metallurgist, Diamond Chain & Mfg. Co., Indianapolis.

I found this paper very interesting. Naturally a study of the effect of one temperature on the fatigue limit of carburized alloy steels raises more questions than it answers. For instance, is the decrease in fatigue endurance limit progressive with increasing temperature from room temperature to 250 degrees Fahr. and what happens above 250 degrees Fahr.? Also what results would be obtained if uniform test pieses without the notch were used?

I was particularly interested by the author's findings that samples containing a large amount of retained austenite showed a greater reduction of endurance limit at 250 degrees Fahr. than those which were treated to produce essentially martensite.

The data in the paper do not include any comparison between the fatigue endurance limits at room temperatures of S.A.E. 2515 treated to produce austenite with those of the same steel reheated to produce essentially martensite. Nor are any given for S.A.E. 3312.

From my experience, I would expect that the austenitic samples would show lower fatigue endurance limits at room temperature for we have found that S.A.E. 4615 and 2315 carburized and treated to produce austenite have lower cross breaking strength at room temperature than parts from the same steel treated to produce essentially martensite when both are tempered at the same temperature of not over 350 degrees Fahr.

I have attributed the lower strength of the steels with highly austenitic cases to locked up residual stress at the extreme surface set up by the difference in thermal expansion of the austenite and martensite. Is it not possible, if this idea is correct, that increasing the operating temperature to 250 degrees Fahr. from room temperature might increase the amount of these locked up stresses and thus lower the endurance limit? Such an increase in residual

stress would be immediately effective in lowering the endurance limit, would not show up as a structural change under the microscope and would take place in spite of any previous tempering operation provided the austenite were not broken down. Further, if this idea is correct, the decrease in endurance limit would probably be progressive with increasing temperatures up to 250 degrees Fahr. and above that temperature until either the martensite or the austenite or both are decomposed. Also, the notch would magnify any stress concentration caused by this differential expansion and thus magnify the effect of increasing the operating temperature.

It is for these reasons, that I raised the questions of the effect on the endurance limit of carburized steels of progressively increasing the temperature and of the use of unnotched test pieces.

Written Discussion: By C. T. Eakin, Westinghouse Electric and Manufacturing Co., East Pittsburgh, Pa.

We are indebted to the author for another very interesting and valuable contribution to the literature on the performance of gear steel. The literature is not at all crowded with papers dealing with the performance of steel at moderately elevated temperatures especially where sliding surfaces and lubricating materials are factors.

The author has brought out in a very clear manner that the notch-bar endurance value of these steels, when tested in rear axle lubricant at 250 degrees Fahr. decreases about 10 to 12 per cent below that at room temperature.

We are wondering if the author has made any attempt to separate the part played by the oil and the part played by temperature by repeating the notch-bar test at 250 degrees Fahr. without oil. Work done by the National Physical Laboratory, England,[1] on Armco iron and steel of 0.17, 0.24, and 0.51 per cent carbon indicates that the fatigue strength of these steels drops as the temperature is changed from zero to 150 degrees Cent. and then increases as the temperature rises to 325 degrees Cent. (615 degrees Fahr.). For instance, on the 0.51 per carbon steel the endurance limit at room temperature is 15 tons and at 150 degrees Cent. (300 degrees Fahr.) it is 13½ tons or a loss of 10 per cent.

From data by Wishart and Lyon[2] in their paper being presented at this convention it is observed that in changing from −40 to +70 degrees Fahr. the endurance value of S.A.E. 1020 decreases 30 per cent, a 0.75 per cent carbon steel about 7 per cent, and a 3 per cent chromium steel about 6 per cent. From the data cited it is noted that when the endurance value is plotted against temperature the curves have a downward slant for temperatures up to about 250 degrees Fahr. It would seem possible that the decrease in endurance found by Mr. Boegehold may be due entirely to the effect of temperature.

Referring to paragraph 2, it is estimated that the temperature at which the surface of the gear teeth operates is about 250 degrees Fahr. We are wondering if the author has in mind the surface material to a depth of say

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[1] H. J. Tapsell, and W. J. Clenshaw, "Properties of Metals at High Temperature," National Physical Laboratory, Department of Scientific and Industrial Research, Engineering Research Special Report No. 2, London, 1927.

[2] H. B. Wishart and S. W. Lyon, A.S.M. Convention, 1936, "Effect of Overload on the Fatigue Properties of Several Steels at Various Low Temperatures."

that of carburization or the temperature of the outer atomic layer where the teeth make contact. It is evident that practically all of the heat is liberated at the surface atoms of the contact and must be absorbed by the backing up metal and the lubricating oil. Factors affecting the temperature of these atomic surfaces are pressure, speed, coefficient of friction and thermal conductivity of the steel. It is hardly to be expected that the tooth surfaces form absolute fits. There must, therefore, be some high points that take more than their share of the load. At these places there will be a great deal more heat liberated. It is the writer's opinion that frequently surface temperatures far in excess of 250 degrees Fahr. are developed momentarily on the tooth contact surfaces.

For example, in their work on the physical properties of sliding surfaces, Bowden and Ridler[3] found that even with polished surfaces well lubricated and running smoothly with a low coefficient of friction, a surface temperature of 600 degrees Cent. (1110 degrees Fahr.) was obtained.

The high temperature of the surface atoms at the point of contact may account for some of the discrepancies between calculated stress and service performance reported in the paper by Alman and Boegehold mentioned in the first paragraph of the paper.

Written Discussion: By H. W. McQuaid, Republic Steel Corporation, Massillon, Ohio.

This paper represents another addition to our data covering the fatigue values of carburized steels. It indicates, in my opinion, that the internal stresses on the surface of a carburized and quenched steel are increased over those at room temperature when the part is heated to a temperature of 250 degrees Fahr.

This increase in internal stress is due to the structural changes involved on heating unstable austenite or martensite for a long period of time. This change in structure is difficult to see under a microscope, but is represented usually by a change in the crystal structure and an increased precipitation of carbides. The change in structure after a long hold at 250 degrees Fahr. is often seen after light etching on the steels used in this test.

It is interesting to note that the higher alloyed steels do not show fatigue properties when tested at 250 degrees Fahr. which would warrant their use in preference to the lower alloyed steels. It is also interesting to note that they indicate a rather definite advantage of the S.A.E. 4615 over the S.A.E. 3115. In this case the S.A.E. 3115 was lower in nickel and manganese than the preferred S.A.E. 3100 gear analysis. It is believed that under the usual conditions there is little to choose between these two steels on a test such as this.

In the summary it is indicated that the difference in the fatigue life of the steels tested at the higher temperatures as compared to room temperature is an indication that gears made from these steels will show a corresponding decrease in fatigue life when tested at the higher temperature. This is quite possibly true, although it is believed that the skin temperature developed in a heavily loaded gear would be considerably higher than the 250 degrees Fahr. used in this test and hence the difference may be even greater between the gear life in service and the gear life at room temperature. It is believed however, that

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[3]Bowden and Ridler, "Physical Properties of Surfaces, Part III," *Proceedings*, Royal Societies, May 1, 1936, p. 640-656.

those properties of resistance to wear and maintained accuracy of contact are of greater importance than fatigue strength of the gear surface alone. In my opinion, the fact that the fatigue life is very greatly reduced by an increase in temperature to 250 degrees Fahr. is of considerable importance and it would be especially true in carburized shafts operating at such a temperature. Where the primary requisite, however, is one of surface resistance to wear, we are inclined to believe that the application of this data is of uncertain value. It indicates that there is great advantage to be gained in keeping as low a temperature as possible on heavily loaded carburized parts, especially such parts as roller bearings or gears which are so designed that they operate very close to the theoretically designed position.

The results obtained in this paper would indicate also that a double treated gear is to be preferred to one quenched directly from the carburizing box and yet we have many instances in actual practice where the reverse is true. We know that the presence of austenite in the surface zone of the carburized and quenched part will reduce the strength by increasing the internal stresses. In some parts it is necessary to reduce the austenite in the surface zone by a double treatment. This is particularly true of the highly alloyed nickel and nickel-chromium steels, but the increase in strength obtained by double treatment is obtained at a sacrifice of resistance to abrasion which is obtained by the work hardening of the austenitic structure. This improvement in the surface resistance to scuffing, obtained by work hardening of the austenite, is a very important factor in the satisfactory performance of rear axle drive gears.

It is noted on page 247 that the 100-degree notch and 0.025 inch radius is indicated as having only slightly less concentration of stress than would be obtained with a 30-degree notch and a 0.010 inch radius. While this is shown to be the case as far as the rotating beam fatigue test is concerned, we have found that it does not check with the results obtained when using an Izod Impact Test as a basis of comparison. This test indicates a rather large difference in stress concentration between the 30-degree notch and the 100-degree notch.

Author's Closure

I agree with Mr. Focke that the amount of information presented in this paper raises more questions than it answers. That statement may be appropriately made any time a new fact is discovered and reported. One example of this tendency is the great amount of work done and reported on nitriding to answer the many questions that arose when nitriding was first introduced.

In view of the results presented earlier this week by Wishart and Lyon[4] showing a progressive increase in fatigue resistance with decreasing temperatures, it appears logical to expect that the decrease in fatigue resistance is progressive up to 250 degrees Fahr. and above that temperature until a temperature is reached at which tempering of the quenched structure begins when there may be some change in fatigue resistance depending upon the nature of the initial structure.

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[4]H. B. Wishart and S. W. Lyon, "Effect of Overload on Fatigue Properties of Several Steels at Various Low Temperatures," Preprint No. 32.

It is presumed that plain specimens would also show the influence of temperature similar to the notched specimens for note that Wishart and Lyon used plain specimens. The comparison between a martensitic structure at room and elevated temperature should be made and we intend to. We would expect the effect of temperature to be much less marked than upon the austenitic structure judging from tests made at room temperature on S.A.E. 3115 steel in the martensitic condition which showed a fatigue resistance no higher than that possessed by S.A.E. 2515 in the martensitic condition at 250 degrees Fahr. I am not able to offer any comment as to the correctness of Mr. Focke's theory that residual stresses in the austenite explain its lower fatigue resisting properties.

Mr. Eakin's suggestion that the loss of resistance to fatigue may be entirely a result of temperature may be the right answer. The data of Wishart and Lyon lends credence to this thought.

The author does not regard as correct, however, Mr. Eakin's second concept that temperatures far in excess of 250 degrees Fahr. occur in the outer atomic layer of the gear tooth because if temperatures only slightly above 300 degrees Fahr. were present continually or repeatedly for a great enough number of times, the hardness of the gear tooth surface would be reduced by tempering. We find that gear teeth after service in passenger cars are still file hard—evidence that temperatures have not been much over 250 degrees Fahr. In truck or bus gears temperatures are sometimes high enough to cause softening of the gear tooth, as is suggested in discussion by Mr. McQuaid.

It is not intended to imply as the results of tests reported in this paper that gears should be double treated to produce martensite simply because fatigue specimens in this condition showed greater life. We appreciate only too well that the steel and heat treatment we select for gears must always be a compromise determined by several conflicting influences. Three of these influences which are of prime importance are: resistance to repeated stress application at notches at high temperature, absence of distortion in heat treating and resistance to wear. It so happens that the treatment which gives the highest resistance to fatigue is the one which causes most distortion and consequently additional stress concentration which cancels out any higher strength advantage obtained by that heat treatment. The only purpose of this paper is to relate the interesting fact that resistance to fatigue is adversely affected by temperatures above atmospheric.

PHYSICAL PROPERTIES OF AXLE SHAFTS

By H. B. Knowlton

Abstract

This paper covers an investigation made to determine the best type of steel and the best combination of physical properties for certain tractor axle shafts. While the investigation is limited to tractor axles, the fundamental principles will probably apply not only to other axles but to many commercial articles.

The correlation between many types of laboratory tests and practical usefulness of axles is discussed. The following types of tests are discussed:—static tension, static torsion, tensile impact, notched bar impact, repeated impact, rotating beam fatigue, torsion fatigue, torsion impact on small specimens, and torsion impact on production axles.

The effect of design and service conditions upon the selection of steel and the physical specifications are considered.

The relationship between the different types of steel and the optimum physical properties is discussed. Particular attention is given to the effect of the penetration of hardness and other inherent properties of different steels.

FOR several years the writer's laboratory has been making tests of axles and axle steels in an effort to determine the best type of steel and the best combination of physical properties. This problem is probably more serious in the case of tractors than in the case of pleasure cars, due to the fact that the tractors are operated during the major portion of their life in low gear with almost full motor torque, while pleasure cars encounter such conditions only occasionally. Futhermore, in the case of farm tractors having wheels equipped with spade lugs, there is always a possibility of one lug slipping and the next engaging with a severe shock. It must also be remembered that the torque on tractor axles is high in proportion to the horse power of the motor due to the low speed of operation.

A paper presented before the Eighteenth Annual Convention of the Society held in Cleveland, October 19 to 23, 1936. The author, H. B. Knowlton, is metallurgical engineer, Gas Power Engineering Department, International Harvester Co., Chicago. Manuscript received June 24, 1936.

For equivalent horse power the torque varies inversely with the speed.

There may have been a time when tractors could be designed many times as heavy as necessary, but such a practice would be unthinkable under the present economic conditions. The problem of the engineering metallurgist is to select the steel and specify the physical properties, which will permit the most economical design with assurance of long life.

Service Tests

The best test is actual service, and while all new model tractors are given service tests, it is not advisable to depend solely upon such tests for the selection of steels and physical properties for axles. Considering that tractors are expected to withstand severe service for at least 7 to 10 years, it would seem that service tests would necessarily be long and expensive. It is not quite as bad as it seems, because an axle which is considerably over-stressed will probably fail in a relatively short time. Five million revolutions of successful performance may be considered as fair assurance of indefinite life.

It is difficult to make an accelerated service test because full motor torque and low gear operation is considered as a normal condition. Placing a larger motor in a small tractor is not practicable because too many parts other than the axles would be overstressed. Another disadvantage of service tests is the extreme difficulty of keeping all conditions constant during a long test.

Service Records

Records of service failures, if complete, should give valuable information. Fortunately, however, failures have not been numerous and only a little information can be gained from this source. When service failures do occur, the customer seldom gives an accurate statement of the service performed. Several significant points are evident from the few service failures which have been examined.

While industrial axles are sometimes so severely over-loaded that bending type of failures are produced, axles of farm tractors generally fail in torsion at the small spline at the differential end.

No actual service failures are available at the time this is being written. A review of the returned goods reports of the last five

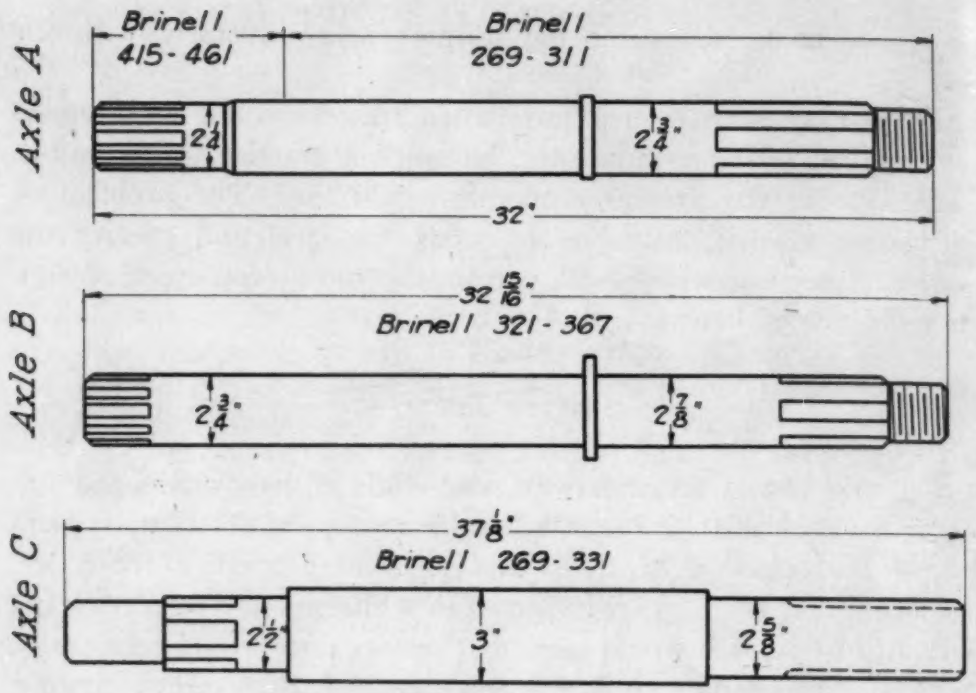

Fig. 1—Designs of Typical Tractor Axles.

years, shows that some axles failed by a smooth shear in the small spline similar to that shown in Fig. 2; some shatter producing ruptures similar to Fig. 3, and at least one axle was found to have fatigue cracks extending inward from the roots of the splines.

Another important point learned from service failures is that the failure occurs without any apparent permanent twist of the axle. In other words, the failure is probably produced either by impact or by torsion fatigue.

Choice of Physical Properties

The problem of improving the physical properties of axles may be approached in four ways; first, by changing the hardness and consequently the other physical properties of a given steel; second, by changing the steel; third, by changing both; and fourth, by changing the design of the axle. All four have been tried and will be considered in order.

The steel employed in production was 3140 with a 6-8 grain size. The surface hardness specified varied somewhat with different axles. The designs and physical specifications of three typical axles are shown in Fig. 1. The first problem attempted, was to determine what hardness range would produce the most favorable

combination of physical properties for each axle using this type of steel.

We will consider first the information which may be obtained from the ordinary laboratory tests.

Hardness Test—The hardness test, either Brinell or Rockwell, is the most commonly used of any of the physical tests; this being about the only type of nondestructive test which gives any indication as to the other physical properties of the steel.

It is perfectly feasible to require that 100 per cent of the parts heat treated be submitted to a routine Brinell test on the surface. Fortunately, there is a definite relation between Brinell hardness and tensile strength. In the case of axles, however, the penetration of hardness is just as important as the surface hardness. This can be determined only by destroying the axle and making hardness tests on the cross section.

Tensile Test—The tensile test is ordinarily performed on a standard 0.505-inch diameter 2-inch gage length test specimen. The properties determined are: (1) tensile strength; (2) yield point, proportional limit, or elastic limit; (3) elongation; (4) reduction of area. To this may be added the total energy consumed in breaking the specimen and sometimes other properties.

All of these properties depend upon the exact design of the specimen. A change of design will produce a considerable variation in the results. For example, a notch in a tensile specimen may produce as much as 64 per cent increase in the tensile strength with even a greater decrease in the elongation and reduction of area. (1)[1] In correlating tensile properties with expected properties of any given article such as an axle or a gear, the engineer must decide whether the design of the article will produce a variation in the properties. In most cases it will.

The elongation and the reduction of area are frequently taken as a measure of the toughness of the steel. This, however, is not always reliable. When certain steels are tempered in the "blue brittle range" (about 500-700 degrees Fahr.) the notched bar impact strength drops to a low figure without any decrease in the elongation and reduction of area. When tempered at higher temperatures there is a great increase in the impact strength, and only a slight increase in the elongation and reduction.

Another important point to be decided is the location from where

[1]The figures appearing in parentheses refer to the bibliography appended to this paper.

the tensile specimen should be taken. It is quite customary to cut the specimen midway between the center and the surface. While this may represent the average properties, it must be remembered that the maximum stress occurs in the outer fibres of an axle.

Notched Bar Impact Test—In spite of its many critics, this test does give some valuable information not derived from tensile, torsion, or fatigue tests. No attempt will be made to give a complete discussion of this test. Hoyt (2) has recently presented an excellent treatise on this subject. He brings out that this test excells in detecting brittleness due to over-heating, temper brittleness, presence of inclusions, harmful types of structure, etc. In many cases the tensile test or even microscopic examination fails to detect the brittleness.

Hoyt (2) also shows that the presence of the notch entirely changes the type of stress occurring in the specimen, and tends to produce cleavage and prevent plastic flow. The present writer wishes to call particular attention to the latter point. It is this point which makes it so difficult for the metallurgist to interpret the notched bar test correctly in the selection of the best type of steel and the best range of hardness for axle shafts or other production parts, which involve only a small degree of the "notch effect."

Impact Tensile Test—The impact tensile test, using a specimen such as shown in Fig. 9, has proven very valuable in determining the resistance of bolts and studs to axial impact stresses. In this type of test there is not the concentration of stress that occurs in the notched bar impact test. Just how this correlates with axle problems will be discussed later.

Repeated Impact Test—A modification of the Stanton test was employed in this investigation. Three types of specimens were used (see Fig. 9). In each test, the specimen was subjected to a series of blows of the same magnitude. The specimen was rotated 90 degrees between blows, and the number of blows necessary to produce failure were recorded. This is a modification of the regular Stanton procedure but it is believed that it gives some valuable results in showing the effect of design on the optimum hardness.

Static Torsion Test—The static torsion test is a quick convenient test to apply to finished axles. It is particularly valuable in the routine testing of production axles. Errors in steel or heat treatment producing poor penetration of hardness are quite evident.

Two types of fractures produced by static torsion are shown in

Figs. 2 and 3. Of these the shear type is the more common. The shatter type will be discussed later. It will be noted that neither of these fractures is similar to the torsion fatigue type shown in Fig. 5.

Rotating Beam Fatigue Test—In this test a bending load is applied to a rotating specimen usually similar in design to that shown in Fig. 9. It will be noted that this specimen is machined to

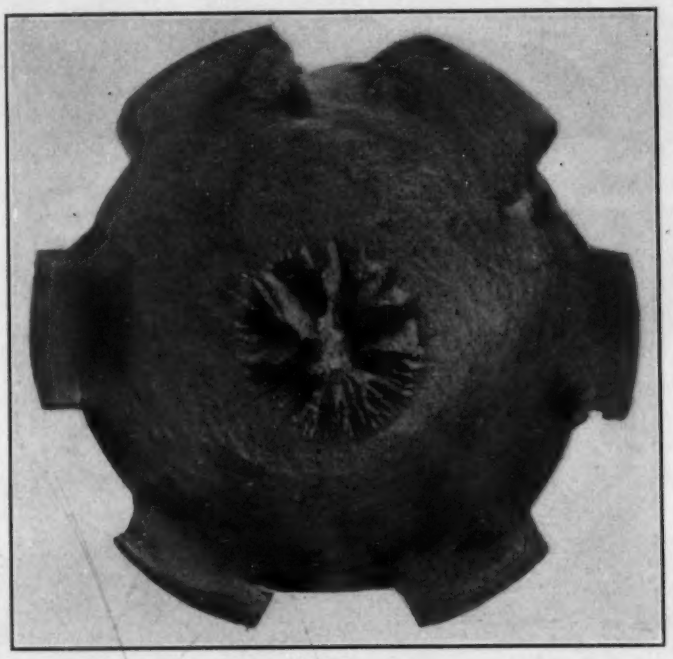

Fig. 2—Static Torsion Fracture—Shear.

a very large radius (9.85 inches). Consequently, there are no concentrations of stress involved in this test, such as occur at the roots of the splines of an axle shaft, or in the fillets of many other production articles.

In making fatigue tests it is customary to run a series of specimens at reducing magnitudes of stress and plot the stress against the number of cycles forming an S-N curve. In the case of steel the curve becomes fairly flat after several million cycles of stress. It is usually assumed that a stress which does not cause failure within 10 million cycles can be withstood indefinitely; consequently, such a stress is commonly considered as the endurance limit in fatigue. In the present investigation the S-N curves were considered sufficiently flat for estimating purposes after 2-5 million cycles of stress.

In applying this test to engineering problems, it is frequently more important to determine the stress which can be withstood for

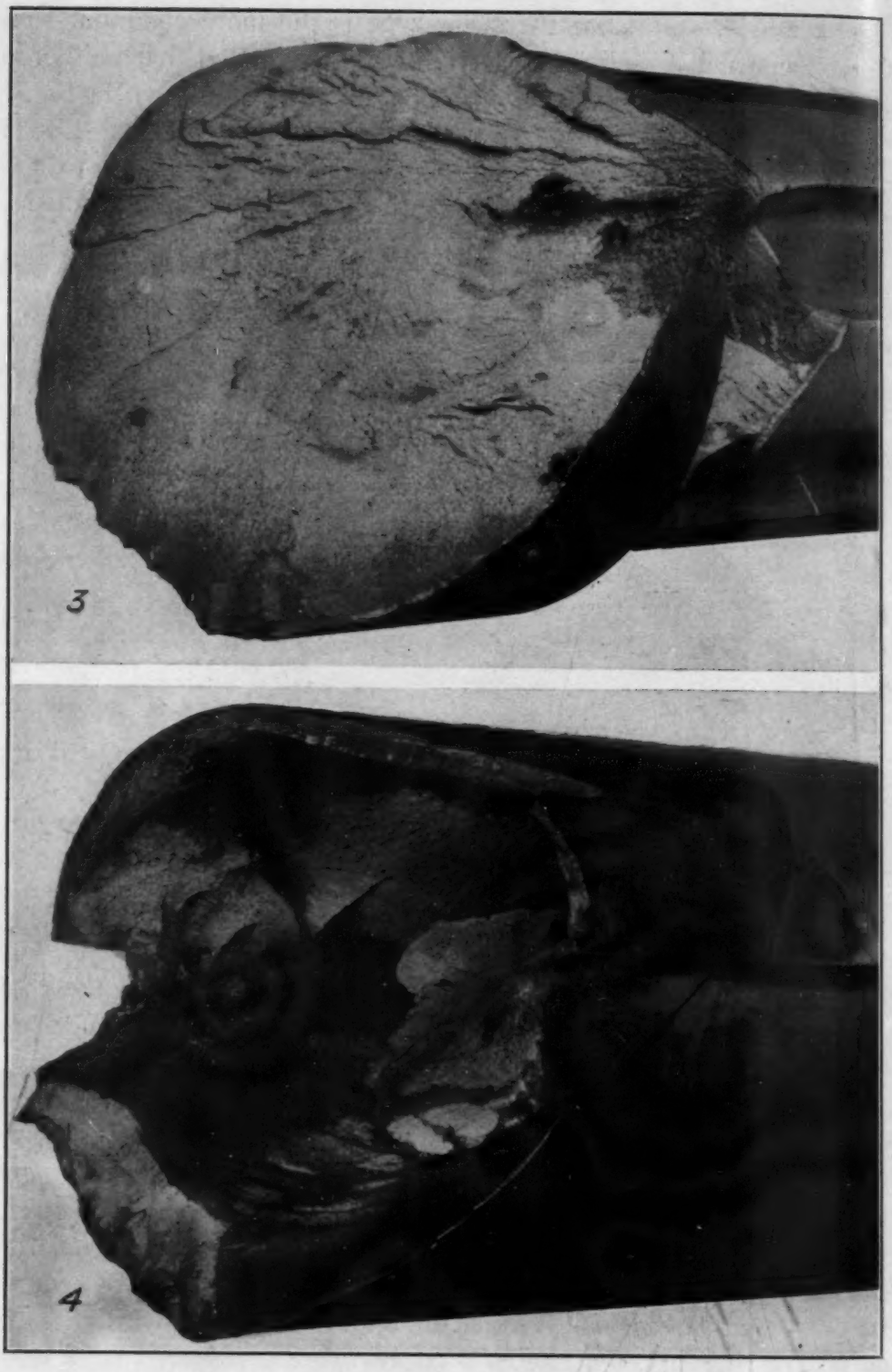

Fig. 3—Static Torsion Fracture—Shatter.
Fig. 4—Torsion Fatigue Fracture—Large End. (Dark Areas Are Planes of Fatigue; Light Areas Final Fracture).

a number of cycles corresponding to the expected maximum service of the production article under consideration. For example, the axle shaft of a pleasure car during the entire life of the car probably does not encounter even 100,000 cycles of the maximum stress, which occurs only under conditions of full motor torque and low gear operation.

French (3) has shown that applying a stress above the endurance limit for a relatively small number of cycles may even raise the endurance limit. He has shown the value of plotting a "damage line" on the S-N diagram, which shows the number of cycles of stresses of different magnitude which can be withstood without reducing the endurance limit.

In the case of tractor axles, however, it is probably the actual endurance limit, which is of the most importance, because such axles may actually encounter millions of cycles of low gear operation with almost full torque of the motor applied.

The principal difficulty in applying the results of the rotating beam fatigue test to the study of axles, gears, and other production parts lies in the fact that such parts must necessarily have fillets with radii much smaller than those of the test specimen. As a result the endurance limit as determined by the rotating beam test is frequently much higher than that obtained by performing a fatigue test on the actual production article (assuming that the ordinary methods for figuring stresses are employed). The rotating beam endurance limit may be said to be a figure which may be approached but seldom reached by production articles.

Boegehold (4) experimented with fatigue specimens of different designs in an effort to make a laboratory test, which would correlate with the service performance of gears. The test specimen which gave the best correlation contained a groove producing about the same concentration of stress as that which occurs at the roots of the gear teeth under consideration.

Torsion Fatigue Test—In the writer's laboratory a special type of fatigue test has been employed for axles. In this test one end of the axle is held stationary, while a torsional load is alternately applied to and released from the other end of the axle by means of a long lever arm actuated by an eccentric. The load is applied to the splines by means of a chuck in the lever arm. In other words, it is attempted to simulate service conditions as nearly as possible in applying the load to the axle. As a regular production design axle

is used as a test specimen, the concentration of stress should be the same as it is in service. The stress is varied from almost zero to maximum. The only way in which this test does not simulate service conditions is in the speed of application of stress, which may be somewhat low.

Merit Indices—It need hardly be stated that the hardness which gives the highest figures on one type of test will probably give lower figures on another. In general it may be stated that strength increases with hardness, but toughness decreases. At just what hardness the best combination of strength and toughness occurs, is not so easy to decide. There have been numerous mathematical formulas proposed, which give quality factors or merit indices. Among these may be mentioned the following:—(5)

1. Tensile Strength × Elongation.
2. Elastic Limit × Reduction of Area.
3. Elastic Limit × Elongation.
4. Elastic Limit × $\sqrt{\text{Elongation} \times \text{Reduction of Area.}}$
5. Elastic Limit × $\sqrt[3]{\text{Elongation} \times \text{Reduction of Area.}}$
6. Tensile Strength × Reduction of Area.
7. Tensile Strength × Izod Impact Value.
8. Elastic Limit × Izod Impact Value.
9. (6) $\dfrac{\frac{1}{2}\,(\text{Tensile Strength} + \text{Elastic Limit}) \times \text{Elongation}}{100 - \text{Reduction of Area.}}$

10. (7) $$\frac{w}{W} = \sqrt[3]{\frac{y}{Y}}\ \sqrt[2]{\frac{u}{U}}\ \sqrt[3]{\frac{e}{El.}}\ \sqrt[3]{\frac{r}{R}}$$

$$R = \frac{100\ R.A.}{100 - R.A.}$$

11. (8) $$n = \sqrt[2]{\frac{(T - Y)\ (E + R)}{1000}}$$

These formulas, however, seem to be unsatisfactory for determining the optimum hardness of any given part, or even for determining the best type of material to be used. Most of the formulas include either elongation or reduction of area as a factor. This may give an undue importance to plasticity. Using some of these formulas it may be found that an extremely plastic material has a higher merit index than a heat treated alloy steel. As will be discussed later, it is the writer's thought that the useful function of plasticity is to permit sufficient deformation of the more highly stressed fibres, in a nonuniformly stressed article to bring about a more favorable distribution of the stress. If this is the case there

is a useful limit of plasticity depending upon the design of the article and the type of service which it must withstand. A higher degree of plasticity would not increase the merit of the steel so far as that particular application is concerned.

Obviously, no matter which formula is used, there will only be one hardness for a given kind of steel which will give the highest merit index. Apparently then this hardness should be used regardless of whether the article to be made is a steering knuckle, an axle, or a part subjected only to static tensile stresses. It is the writer's belief, therefore, that none of these formulas are valuable in determining the optimum hardness of a given kind of steel for a specified part.

Test Procedure

In order to determine the best hardness for tractor axle shafts when made of the steel used in production, all of the tests just described were performed on S.A.E. 3140 steel. With the exception of the fatigue tests, all tests were made in duplicate or triplicate. Additional tests as later described were also employed.

While from a theoretical standpoint it might be desirable to cut all of the test specimens from axles made from a single heat of steel, this was not considered practicable on account of the time and expense involved. The static torsion and torsion fatigue tests were

Table 1
Materials Used for Preliminary Tests—S.A.E. 3140 Steel

Test	Size	Bar	Chemical Analysis C	Mn	S	P	Ni	Cr	Grain Size and Normality
Fatigue—	⅝" Rd.	1	0.38	0.72	0.021	0.013	1.28	0.61	4– N
(Rotating Beam)	⅝" Rd.	2	0.40	0.75	0.024	0.013	1.27	0.62	4– N
	⅝" Rd.	3	0.38	0.74	0.026	0.021	1.25	0.64	4– N
Tensile (Threadless)	1" Rd.	6	0.38	0.71	0.026	0.027	1.23	0.64	7– N
Repeated Impact	⅞" Rd.	4	0.40	0.63	0.022	0.013	1.24	0.64	4– N
(⅛" Radius)	⅞" Rd.	5	0.40	0.62	0.026	0.016	1.24	0.64	3–4 N
Repeated Impact	⅞" Rd.	1	0.38	0.62	0.033	0.022	1.31	0.51	*5–6 N
("V" Notch—¼" Radius)	⅞" Rd.	0	0.38	0.55	0.030	0.021	1.36	0.54	5–6 N
Izod (Tensile)	⅞" Rd.	–	0.35	0.54	0.019	0.019	1.19	0.66	–
Izod (Notched Bar)	⅝" Rd.	A	0.41	0.65	0.018	0.022	1.20	0.71	7– N
	⅝" Rd.	B	0.41	0.66	0.018	0.023	1.21	0.73	7– N
	⅝" Rd.	C	0.38	0.67	0.018	0.024	1.20	0.73	7– N
	⅝" Rd.	D	0.41	0.64	0.019	0.023	1.20	0.73	7– N
	⅝" Rd.	E	0.41	0.65	0.018	0.023	1.15	0.69	7– N
	⅝" Rd.	F	0.41	0.65	0.018	0.022	1.22	0.73	7– N

*Some large grains at surface.
N = Normal.

made on regular axles. (See Fig. 1-A). The other specimens were made from small heat treated bars. Each set of test specimens was made from a single heat of steel. The chemical analysis and McQuaid-Ehn grain size of each heat of steel used is shown in Table I.

Results—The results of the rotating beam fatigue and the torsion fatigue tests are shown graphically in Figs. 10 and 11. The endurance limits were computed from these graphs.

In order to get a better picture of the correlation of the results of different types of tests, all of the data obtained in the earlier tests is plotted on a single diagram. (See Fig. 12). It will be noted that all of the other physical properties are plotted in terms of Brinell hardness, because the objective of this part of the investigation was to determine the optimum Brinell hardness of 3140 steel for tractor axles.

Discussion of Results

The following generalizations can be drawn from this chart:—The tensile strength and proportional limit increase with Brinell hardness, while the elongation and reduction of area decrease. The total energy consumed in breaking the tensile specimens seems to increase with Brinell hardness above 400 Brinell. The Izod notched bar impact strength decreases with increase of Brinell hardness up to about 400 Brinell. The endurance limit as determined by the rotating beam test also seems to increase with Brinell hardness, although the curve is fairly flat. Torsion fatigue test, on the other hand, shows a maximum endurance limit at about 415 Brinell and only a slightly lower figure at 495.

The first series of repeated impact tests were run on specimens having ⅛ inch radius groove. This showed a decided peak at 415 Brinell. From these results it seems that the optimum hardness depended upon the degree of concentration of stress which was encountered.

Conclusions from the First Series of Tests—From these tests the following tentative conclusions were drawn.

(1) Strength increases and plasticity decreases with increase of Brinell hardness.

(2) When the load is evenly distributed and the stresses uniform there is little or no need for plasticity. The highest test values are obtained with the highest Brinell hardness. This seems to be true regardless of the velocity of application of the stress, as is

evidenced by the fact that both the static tensile and the impact tensile tests show maximum values at the highest hardness.

The statement might well be qualified by saying that the maximum values are obtained at the highest hardness, providing other conditions are equal. Large grain size, inherent defects in the steel and the presence of quenching strains and probably some other factors might have a deleterious effect.

(3) The function of plasticity seems to be to permit sufficient yielding of the more highly stressed fibres to produce a more favorable distribution of the stress. The most nonuniformly distributed stress occurs in the notched bar impact test. It will be noted that in this test the maximum value was obtained at the lowest hardness, which is accompanied by the highest degree of plasticity.

The repeated impact test using the specimen with the 1/8-inch radius groove produces a nonuniformity of stress intermediate between that of notched impact bar on the one hand, and the tensile and fatigue specimens on the other. As might be expected it shows a maximum (in terms of number of blows before failure) at an intermediate hardness.

There is undoubtedly some concentration of stress at the roots of the splines of the production axles. It would be expected that this concentration of stress would be intermediate between that of the tensile specimen and the Izod notched bar; consequently, it would be expected that the optimum hardness would also be at some intermediate figure. Results of the torsion fatigue test have indicated that the optimum hardness was probably in the neighborhood of 415-495 Brinell.

In order to secure more data either to confirm or to refute the idea that optimum hardness is a function of the concentration of stress, another series of repeated impact bars were made with a 1/4-inch radius and two sets with 1/32-inch radius. As was expected the peak on the repeated impact curve with the large radius came at a higher hardness, and with the smaller radius at a lower hardness than that already determined with the specimens having the 1/8-inch radius groove. The difference in hardness at which these peaks appeared was comparatively slight. As all tests were run in duplicate, fair checks were obtained, it was thought that even this slight difference substantiated the principle.

Two sets of specimens with the 1/32-inch radius "V" notch were hardened at different temperatures in an effort to determine whether

the final grain size of the specimen had any effect on the optimum hardness. However, the McQuaid-Ehn grain size was relatively fine and consequently neither heat treatment produced a really coarse-grained specimen. The repeated impact curves, consequently, were almost identical. For this reason only one curve for the 1/32-inch radius specimen is shown. (Fig. 12.)

From all of these tests it would seem that the optimum hardness for this particular axle when made of S.A.E. 3140 fine grain steel is probably about 400-450 Brinell.

This hardness agrees with that at which the maxima on the three repeated impact curves appear. It seems quite probable that the concentration of stress at the roots of the splines is intermediate between that of the "V" notch and the ¼-inch radius fillet of the repeated impact specimens.

This hardness also agrees well with the optimum indicated by the torsion fatigue curves. One reason for giving particular consideration to torsion fatigue results is that this type of test seems to make the nearest approach to service conditions. It produces failure below the elastic limit.

Torsion Impact Test—In the tests thus far described, no attempt was made to simulate a possible service condition in which the fractures is at least started by a single torsion impact. Consequently, a number of attempts were made to determine the resistance of this steel at varying degrees of hardness to torsion impact.

In this investigation the writer was fortunate in obtaining the co-operation of Messrs. Luerssen and Greene of the Carpenter Steel Co., who performed the torsion impact tests on several sets of specimens, using the test machine and method developed by them for testing tool steel. (9)

The materials and the heat treatments are shown in Table II. Several designs of specimens were employed; these are shown in Fig. 13. The results are shown in Table III and in the graph, Fig. 14.

The first series of tests were made on standard specimens and specimens of a modified design containing a ¼-inch radius groove. The results of these tests are shown by the dotted lines in the graph. In the second series of tests the specimens used were the standard, the ½-inch groove, and the splined designs. The results of these tests are shown by the solid lines on the graph.

All tests were run in duplicate. Referring to Table III, it will

Table II
Analysis and Heat Treatment of Small Torsion Impact Specimens

	Chemical Analysis Type of Specimen			
	First Series	Second Series		
	Standard and Grooved	Standard	Grooved	Splined
Carbon	0.40	0.42	0.42	0.42
Manganese	0.76	0.69	0.70	0.65
Sulphur	0.033	0.029	0.020	0.028
Phosphorus	0.034	0.020	0.022	0.026
Nickel	1.22	1.29	1.31	1.30
Chromium	0.70	0.58	0.58	0.63
McQuaid-Ehn Grain Size	3–4	4	4	4

Heat Treatment

Normalized .....1600 degrees Fahr.
Heated .....1525 degrees Fahr.
Quenched In .....Oil
Drawn 1 Hour .....At Temperature

be noted that remarkably good checks were obtained on the duplicate specimens indicating that there was little error in the testing technique. The writer is willing to grant that the amount of data acquired in these tests is too small to permit the drawing of accurate conclusions. However, considering that reasonably good checks were obtained, it may be well to give some thought to the results of these tests.

It seems to the writer that at least one point is conclusively proved; that the design of the specimen does have a pronounced effect upon the results obtained by the torsion impact test. Other conclusions can be stated only tentatively and are subject to confirmation or the reverse by further tests.

It will be noted that the curves for the two sets of standard impact bars do not check with each other at all. This may be due to inherent differences in the two heats of steel, although the chemical analyses and grain sizes are similar. The curve for the first set of standard specimens does not show a peak at 250 degrees Fahr. because no drawing temperature below 350 degrees Fahr. was used. The peak at 250 degrees Fahr. on the second set of standard specimens probably indicates that there is considerable relief of quenching strains without decrease of hardness or change of structure at that temperature. Both sets of specimens show minima in torsional impact strength, probably due to temper brittleness in the neighborhood of 500 degrees Fahr. although the exact temperature and the hardness at which the minimum appears is not the same for the two sets of specimens. Also, both curves show secondary maxima at higher drawing temperatures.

Table III
Torsion Impact Results—Small Specimens

Drawing Temperature (°F.)	First Series				Second Series					
	Standard		¼" Groove		Standard		½" Groove		Spline	
	Rockwell "C"	Impact in Ft.Lb.	Rockwell "C"	Impact in Ft.Lb.	Rockwell "C"	Impact in Ft.Lb.	Rockwell "C"	Impact in Ft.Lb.	Rockwell "C"	Impact in Ft.Lb.
None	50.5	110	..	..	56–57	137	56	39	53–55	199
	50	113	..	..	55–56	127	56–57	35	54–56	201
210	..	..	..	..	55–56	137	...	..	...	..
	..	..	..	..	55–56	129	...	..	...	..
250	..	..	..	..	55–56	186	56–57	24	...	..
	..	..	..	..	55–56	179	56–57	32	...	..
275	..	..	..	..	55–56	166	...	..	...	..
	..	..	..	..	55–56	174	...	..	...	..
300	..	..	..	..	55–56	119	55–56	29	52–53	201
	..	..	..	..	55–56	129	55–56	29	52–53	190
325	..	..	..	..	55–56	144	...	..	...	..
	..	..	..	..	55–56	92	...	..	...	..
350	50.5	87	51.5	23	...	..	...	..	...	..
	51.5	105	51.5	21	...	..	...	..	...	..
375	..	..	..	..	55–56	119	55–56	29	...	..
	..	..	..	..	54–56	119	55–56	27	...	..
500	..	..	..	..	51	88	51–52	25	...	..
	..	..	..	..	50–52	85	51–52	22	...	..
550	48.5	77	48.5	23	...	..	...	..	...	..
	47.5	74	48.5	32	...	..	...	..	...	..
625	..	..	..	..	48–49	223	48–49	45	48–49	147
	..	..	..	..	48–49	230	48–50	36	49–50	161
700	46.5	80	46.5	26	...	..	...	..	...	..
	46.5	77	45.5	34	...	..	...	..	...	..
750	..	..	..	..	45	168	44–46	58	44–45	141
	..	..	..	..	44–45	187	44–45	56	44–45	149
800	..	..	..	..	...	..	...	..	42–44	152
	..	..	..	..	...	..	...	..	43–44	138
850	42.5	167	44	37	...	..	...	..	43–44	158
	42.5	178	43.5	30	...	..	...	..	43–44	147
900	..	..	..	..	40–41	148	40–41	58	40–42	166
	..	..	..	..	39–40	168	40–41	56	40–41	149
950	..	..	..	..	...	..	...	..	36–38	152
	..	..	..	..	...	..	...	..	36–37	163
1000	..	..	36	40	...	..	...	..	33–35	147
	..	..	34.5	39	...	..	...	..	33–34	144
1050	..	..	..	..	32–34	242	33–34	43	31–33	149
	..	..	..	..	33–35	224	32–33	43	31–33	147
1200	..	..	..	..	...	..	27–29	67	...	..
	..	..	..	..	...	..	27–29	65	...	..

As might be expected, the impact strength of the grooved specimens was lower than that of the standard ones. In the case of the ¼-inch groove specimen there was undoubtedly a higher concentration of stress and possibly a different type of stress than appeared in the standard specimen. This apparently placed a premium on maximum plasticity. The curve shows a steady rise in impact values with increase of drawing temperature. Why there was no evidence of temper brittleness or a blue brittle range such as frequently reported in Izod notched bar tests is not quite clear.

The curve for the ½-inch groove bar is intermediate in shape and position between that of the small groove and the standard specimen. It was hoped that the splined specimen would satisfactor-

ily simulate a splined axle, as the ratio of the size of the splines to the diameter of the specimen was approximately the same. These specimens took a much greater twist during the test than any of the other specimens, and consequently showed the highest figures for torsion impact strength. The lesser need for plasticity of the material with this design of specimen is indicated by the fact that the second peak on the curve occurs at a lower figure than the original strength. It is very much doubted that the stresses produced in the splined specimen simulate those in a splined axle which fails with little or no permanent twist.

Peculiarly three of the five curves show secondary minima at temperatures between 900 and 1050 degrees Fahr., followed by another increase of strength at higher tempering temperatures. No explanation is offered for this. It may be contended that this and in fact all of the other peculiar characteristics of these curves are due to experimental error, or to the natural unreliability of the test. As stated before while the present writer would like to have many times this amount of data before drawing many definite conclusions, he is unwilling to disregard even this amount of data or to consider it entirely meaningless.

Summarizing, it is believed that these tests show that the design of the specimen has a pronounced effect upon the torsion impact strength and upon the optimum drawing temperature and hardness at which the highest values for torsion impact strength will be found. They also show that as the notch effect and the resulting concentration of stress become more severe, there is need for greater plasticity. It is not impossible that some sort of a small torsion impact specimen may be designed, which will accurately simulate the stresses in a splined axle, but the author would hesitate to attempt to correlate any of the results so far obtained with axles.

The writer is not disputing the merit of this test in determining inherent brittleness in at least some types of steel. Personally, the writer believes that the principle of the torsion impact test is sound but wishes to point out that in applying the results of this test to engineering problems, the relation of the design of the specimen to that of the production part must be taken into consideration. In this respect the torsion impact test is similar to all of the other tests discussed.

Torsion Impact Tests on Axles—The next attempt to determine torsion impact strength of axles and axle steels was made on pro-

duction axles. In this test axles of Type B (Fig. 1) were used. In this test the torsion impact was applied to the axle by dropping a 9000-pound hammer on the end of a 15-inch lever, which transmitted the blow through a spline chuck to the small splines of the axle. The large splines were held stationary in a splined chuck at the other end of the fixture. The results of this test are shown in the graphs (Fig. 15).

In these tests the axles were submitted to a series of blows from successively greater heights until the axle failed. Both the maximum height of drop and the energy absorbed in breaking the axle were computed. The latter was accomplished by measuring the velocity of the falling hammer before and after breaking the axle. This work was done by the laboratory of the W. H. Miner Co. in Chicago.

From these results it would seem that the optimum hardness for these axles is 450 Brinell or higher. One point, however, must be taken into consideration, and that is that the speed of the falling hammer never exceeded 4½ miles per hour. The fractures produced more nearly simulated those of static torsion than torsion fatigue or service failures.

It was also noted that even though the axle was broken by a single blow, it always suffered a twist in the large section. In service it is possible for these axles to be submitted to a shock corresponding to a velocity of 18 miles per hour at a distance of 15 inches from the center of the axle. Whether this higher velocity would produce a difference in results was not known, but was considered worthy of investigation. This will be discussed later.

Tests on Different Materials—The discussion so far has been limited to the determination of the optimum hardness of a single steel for one particular axle. Very recently another series of tests has been started involving a number of different types of steel. These included S.A.E. 3140 previously used, S.A.E. 4140, Amola, and three varieties of S.A.E. 1045. Two different ranges of hardness were employed, representing the highest and the lowest ranges of hardness used in production for tractor axles. The newest design of axle as shown in Fig. 1 C was used. The details of material and heat treatment, and the surface hardness produced are shown in Tables IV and V.

Two axles of each material and hardness were given the static torsion test, after which sections were cut for cross sectional Rock-

Table IV
Materials Used for Torsion Tests—Type "C" Axles

Material	Chemical Analysis							McQuaid-Ehn Grain Size
	C	Mn	S	P	Ni	Cr	Mo	
S.A.E. 3140 Steel	0.43	0.72	0.020	0.022	1.17	0.74	...	5–7 N
S.A.E. T-1340	0.44	1.94	0.029	0.030	...	...	...	6–7 N
S.A.E. 4140	0.43	0.77	0.021	0.021	...	1.08	...	7–8 N
"Amola" (60–63 Carbon)	0.66	0.75	0.014	0.018	...	...	0.27	5–7 (Slightly abnormal)
S.A.E. 1045 (Controlled Hardenability)	0.43	0.71	0.030	0.018	0.23	0.10	...	2–5 N
S.A.E. 1045 F. Q.	0.46	0.74	0.032	0.020	...	0.06	...	*2–5 N
S.A.E. 1045 + Chromium	0.44	0.75	0.022	0.019	...	0.27	...	*5–7 N

*Tendency towards duplexing noted in the core.

well hardness tests. The results are shown in Figs. 16 and 17.

Discussion of Results—While it may not be advisable to draw too definite conclusions from such a small amount of data, as different heats of the same kind of steel may not perform exactly alike, there are a few points which are quite plain.

Probably the greatest difference between axles made from different steels and heat treated for the same surface hardness, lies in the penetration of hardness. Referring to Fig. 16, it will be noted that the reduction of hardness from surface to center is much greater in the case of the three 1045 steels than it is in the alloy steels.

In Fig. 17 the results of the static torsion test are plotted against the surface Brinell. If the same type of steel were used throughout, the ultimate strength and the elastic limit would increase with increase of Brinell hardness, while the total degrees of twist would probably decrease. In the graph it has been attempted to plot lines representing the approximate averages for ultimate strength, elastic limit and degrees of twist. By comparing Fig. 16 and 17, it will be noted that in general when the penetration of hardness is high, the figures for elastic limit and ultimate strength are greater than the average. Likewise when the penetration is low, figures for strength and elastic limit are less than the average.

It will be found that in general the degrees of total twist varies inversely with the surface hardness. It might be expected that the penetration of hardness would also show a good correlation with the degrees of twist, that is when there is a deep penetration of hardness, the twist would be below average and vice-versa, but this does not seem to be always true. It also appears quite probable

Table V

Torsion Tests—Type "C" Axles

Axle No.	Material	Quenching		Drawing		Brinell	Fracture	Remarks
		Temp. Deg. Fahr.	Medium	Temp. Deg. Fahr.	Time Hours			
29	S.A.E. 3140	1525	Water	1020	2	301	Shear	Drawn in lead.
30	S.A.E. 3140	1525	Water	1020	2	301	Shear	Drawn in lead.
31	S.A.E. 3140	1550	Oil	675	4	461	Shatter	Quenched in water from draw.
32	S.A.E. 3140	1550	Oil	675	4	461	Shatter	Quenched in water from draw.
33	S.A.E. T-1340	1550	Oil	800	4	363	Shear	———
34	S.A.E. T-1340	1550	Oil	800	4	363	Shear	———
35	S.A.E. T-1340	1550	Oil	600	4	444	Shatter	Quenched in water from draw.
36	S.A.E. T-1340	1550	Oil	600	4	461	Shatter	Quenched in water from draw.
37	S.A.E. 4140	1550	Oil	1100	3½	363	Shear	First drawn at 1000°; then re-drawn at 1100°.
38	S.A.E. 4140	1550	Oil	1100	3½	352	Shear	First drawn at 1000°; then re-drawn at 1100°.
39	S.A.E. 4140	1550	Oil	700	4	495	Helical Cracks*	Quenched in water from draw.
40	S.A.E. 4140	1550	Oil	700	4	477	Helical Cracks*	Quenched in water from draw.
41	Amola	1475	Oil	1100	4	331	Shear	Previously drawn at 950° and 1000°, 4 hours each.
42	Amola	1475	Oil	1100	4	331	Shear	Previously drawn at 950° and 1000°, 4 hours each.
44	Amola	1475	Oil	700	4	495	Shatter	Previously drawn at 600° for 5 hours.
45	S.A.E. 1045 (Controlled Hardenability)	1525	Water	900	3	331	Shear	———
46	S.A.E. 1045 (Controlled Hardenability)	1525	Water	900	3	341	Shear	———
47	S.A.E. 1045 (Controlled Hardenability)	1525	Water	600	5	495	Shear	Previously drawn at 500° for 3¼ hours.
48	S.A.E. 1045 (Controlled Hardenability)	1525	Water	600	5	477	Shatter	Previously drawn at 500° for 3¼ hours.
49	S.A.E. 1045 F. Q.	1525	Water	900	3	341	Shear	———
50	S.A.E. 1045 F. Q.	1525	Water	900	3	341	Shear	———
51	S.A.E. 1045 F. Q.	1525	Water	600	5	477	Shatter	Previously drawn at 500° for 3¼ hours.
52	S.A.E. 1045 F. Q.	1525	Water	600	5	477	Shatter	Previously drawn at 500° for 3¼ hours.
53	S.A.E. 1045 + Chromium	1550	Water	900	4	352	Shear	———
54	S.A.E. 1045 + Chromium	1550	Water	900	4	341	Shear	———
55	S.A.E. 1045 + Chromium	1550	Water	600	4	477	Shatter	———
56	S.A.E. 1045 + Chromium	1550	Water	600	4	495	Shatter	———

*Shatter and helical cracks probably represent the same type of failure.

that there are other inherent properties having an effect upon twist.

For example it will be noted that the axles made of S.A.E. 1045 plus chromium and heat treated for the higher hardness, show a low degree of twist in spite of the relatively shallow penetration of hardness. These axles were evidently quite brittle at this hardness as was also evidenced by the fact that the ultimate strength was only

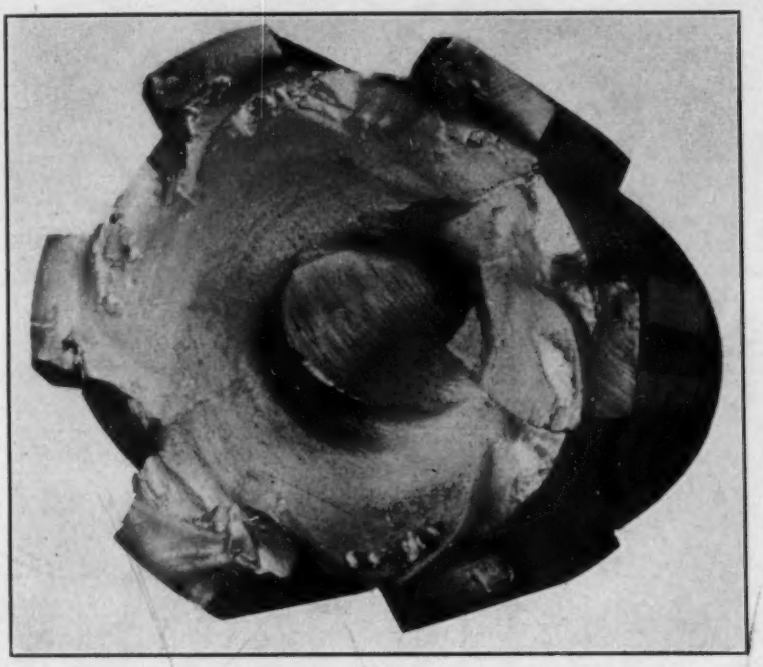

Fig. 5—Torsion Fatigue Fracture—Small End. (Dark Area is a Plane of Fatigue. Light Area is Final Fracture Produced in Static Torsion Machine. Note Cracks extending Inward from Roots of Splines).

slightly higher than the elastic limit. The results of the duplicate axles in this test checked with each other; consequently, it is not believed that the peculiar results are due to experimental error.

On the other hand, the other two axles which were made from the same heat of steel and treated for a lower hardness, showed considerably better than average twist in proportion to hardness with good figures for both elastic limit and ultimate strength. This points to a possible conclusion that there may be a limit to the desirable hardness of this design of axle when made of this type of steel. Much more confirming data will be required before stating this as a definite conclusion, for properties observed may be peculiar to the particular heat rather than to the general type of steel.

It will also be noted that the T-1340 axles showed lower than average figures at the higher hardness but perfectly normal values at

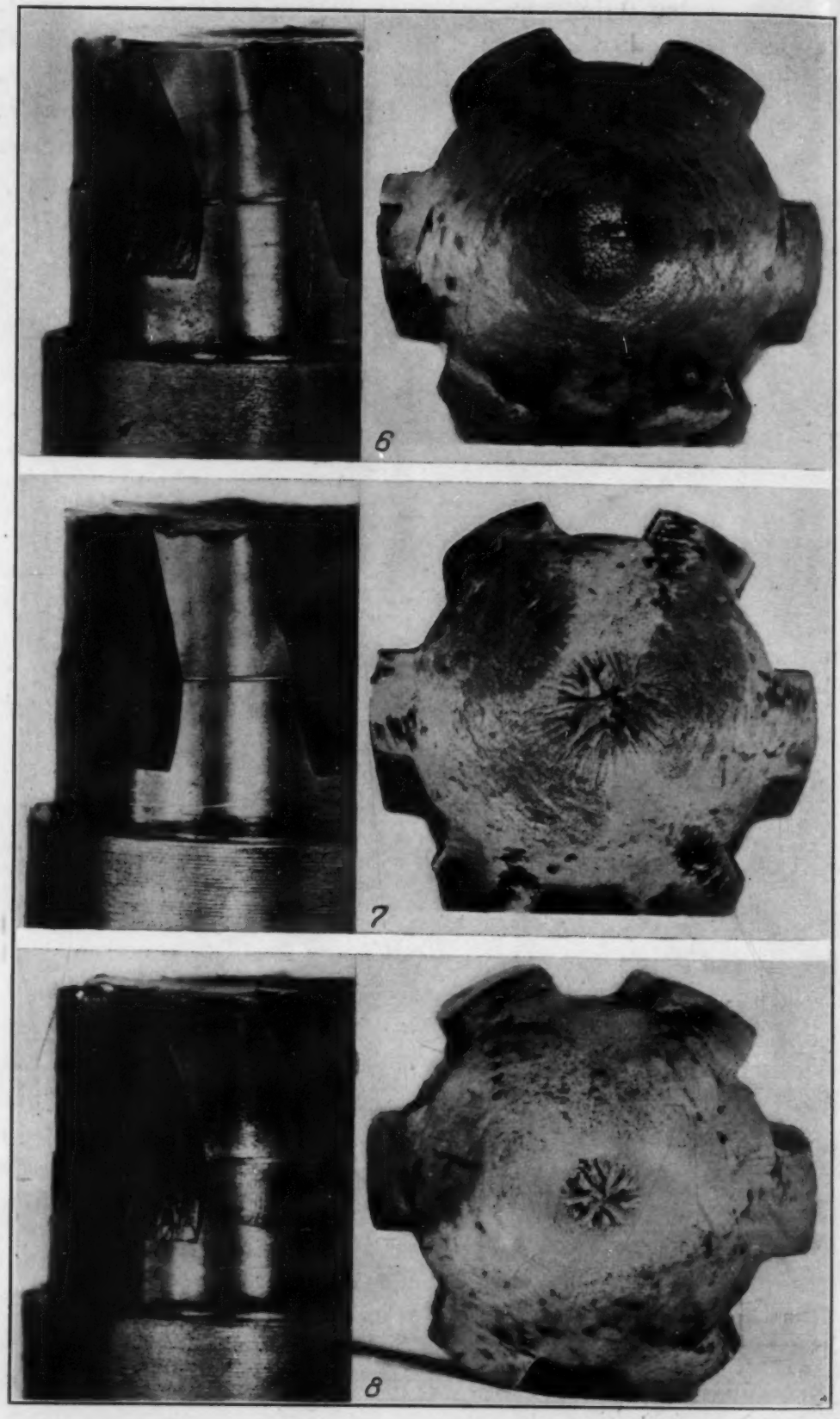

Fig. 6—Torsion Impact Fracture. (S.A.E. 3140 Steel; Low Hardness).
Fig. 7—Torsion Impact Fracture. (S.A.E. 3140 Steel; High Hardness).
Fig. 8—Torsion Impact Fracture. (S.A.E. 1045 F.Q. Steel; Low Hardness).

the lower hardness. It was also found that the T-1340 axles could not be water-quenched without danger of cracking, but that the 3140 axles could be water-quenched without difficulty.

One interesting point noted during these tests was that most of the harder axles broke with the shatter type of fracture (Fig. 3),

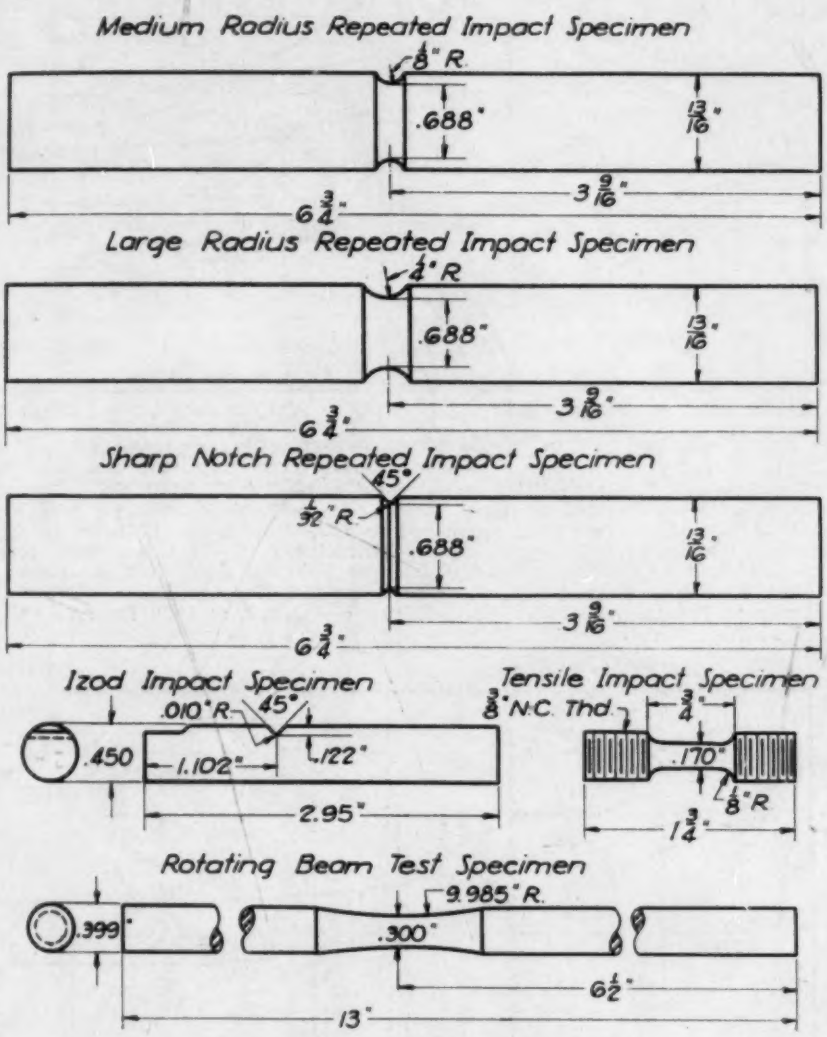

Fig. 9—Designs of Test Specimens.

while the softer axles gave shear fractures (Fig. 2). These were all Type "C" axles (Fig. 1). On the other hand, the type "A" axles, which are made in production of S.A.E. 3140 steel and heat treated for the higher hardness range, so far as the small spline section is concerned, almost invariably gave shear fractures on the torsion test. These axles also showed a permanent twist in the larger and softer section before breaking through the small spline.

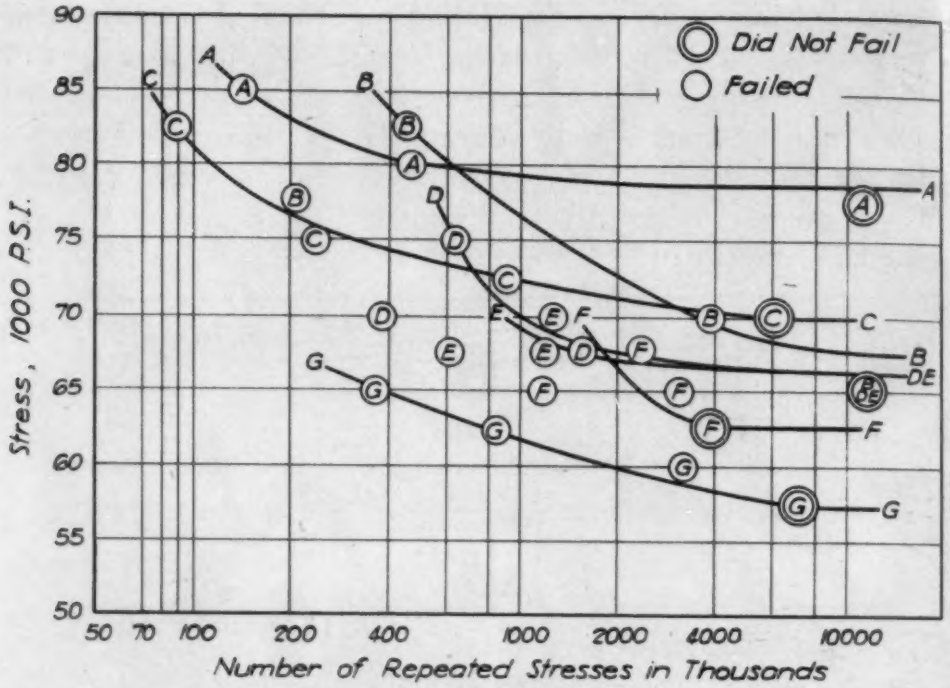

Fig. 10—SN Diagram—Rotating Beam Tests—S.A.E. 3140 Steel. Size of Specimens as Heat Treated 5/8 Inch Round. Normalized at 1600 Degrees Fahr., Heated to 1525 Degrees Fahr., Quenched in Oil, Drawn for 1 Hour as Stated Below. A—Drawn at 400 Degrees Fahr.—Brinell 477, B—Drawn at 550 Degrees Fahr.—Brinell 444, C—Drawn at 650 Degrees Fahr.—Brinell 408, D—Drawn at 750 Degrees Fahr.—Brinell 401, E—Drawn at 850 Degrees Fahr.—Brinell 363, F—Drawn at 950 Degrees Fahr.—Brinell 302, G—Drawn at 1050 Degrees Fahr.—Brinell 272.

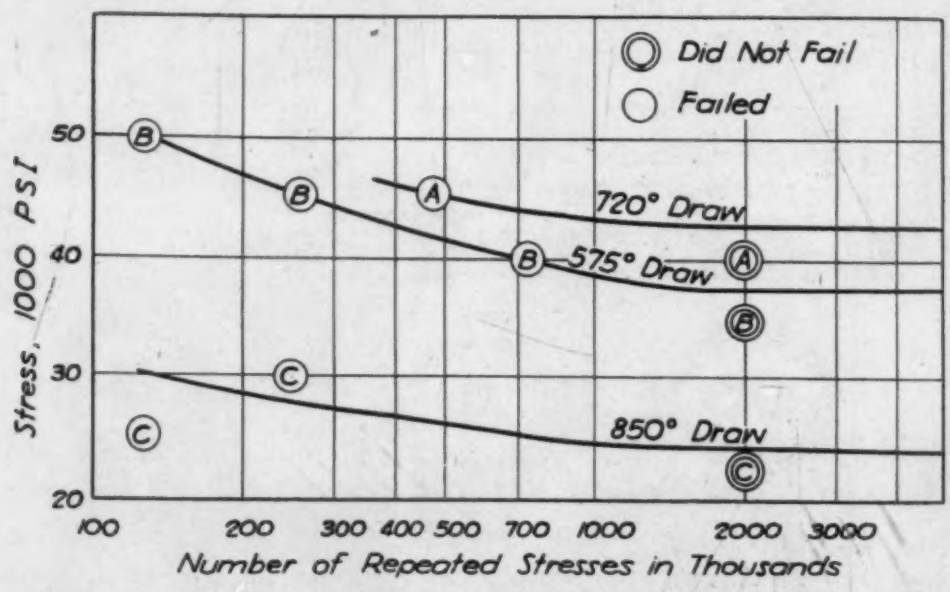

Fig. 11—SN Diagram—Torsion Fatigue Tests of Type "A" Axles. Material—S.A.E. 3140 Steel. For Design See Fig. 1. Heat-Treatment—Normalized at 1600 Degrees Fahr., Heated at 1525 Degrees Fahr., Quenched in Water, Small Splines Drawn 1 Hour as Stated Below: A—Drawn at 720 Degrees Fahr.—Brinell 418, B—Drawn at 575 Degrees Fahr.—Brinell 495, C—Drawn at 850 Degrees Fahr—Brinell 358.

From this it seems that the optimum hardness for the small splines may be higher for the type "A" than for the type "C" design of axle, providing both are made of S.A.E. 3140 steel. Another possible explanation is that the lower hardness of the large section of the type "A" axles protects the spline sections. So far as service failures are concerned, the writer does not consider this probable because most service failures seem to occur at stresses below the elastic limit of the large soft section. If this is true the plasticity of the axle when stressed beyond the elastic limit should not have any effect on the failure.

Torsion Fatigue Tests on Type "C" Axles

Torsion fatigue tests have also been started on the type "C" axles made of different types of steel and heat treated for the high

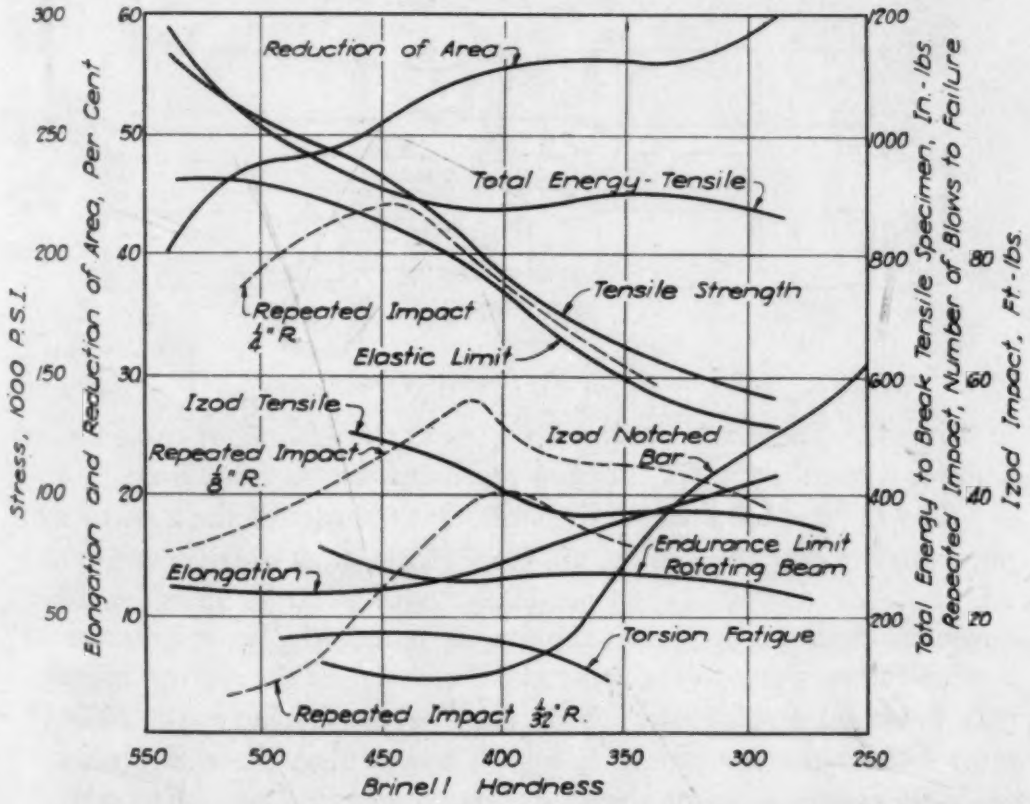

Fig. 12—Physical Properties of S.A.E. 3140 Steel. (Note:—Elastic Limits Were Determined from Stress-Strain Diagrams.)

and low hardness ranges. At the time this is being written only two axles have been broken; consequently, no results can be given.

In making a torsion fatigue test it is customary to discontinue

the test as soon as the first crack becomes apparent. In the present instance, however, it was decided to continue one test until complete destruction of the axle, to determine just how rupture does occur in torsion fatigue. The first fracture appeared at the root of one of the splines at the small end of the axle. The axle finally failed, however, due to a fatigue break starting in the root of the keyway in the large diameter section. The appearance of this fracture is shown

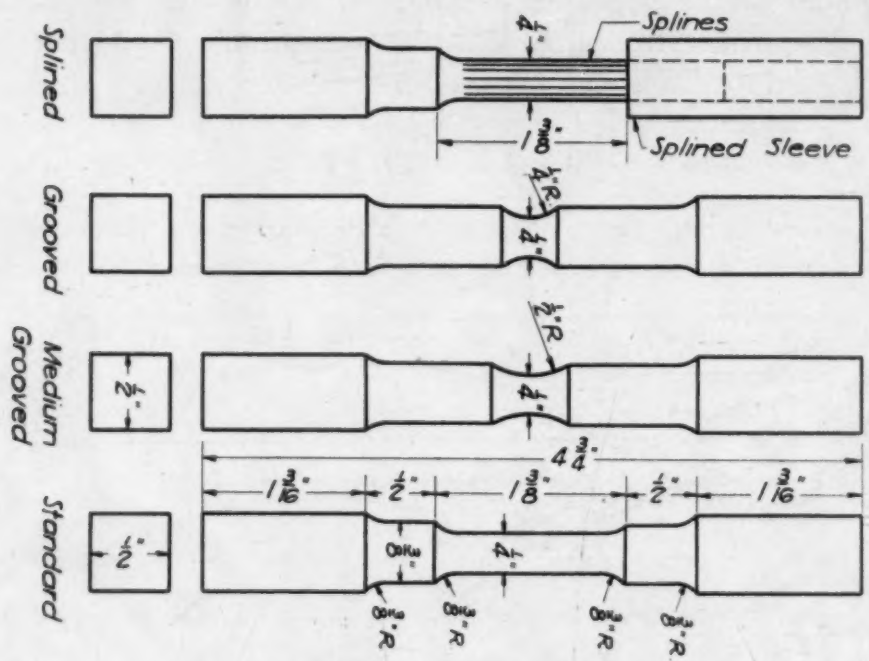

Fig. 13—Torsion Impact Specimens.

in Fig. 4. The dark surfaces represent the plane of progressive failure, while the lighter surfaces represent the final fracture.

It will be noted that fatigue started at the root of the keyway and progressed along the conical plane until the cross section still intact was reduced sufficiently to produce rupture. It is quite possible, however, that this type of rupture is not fairly representative of service failures due to the fact that the chuck in the testing machine may have delivered too much of the load to the keyways, while in service the hubs are clamped in such a way that the round portion probably carries most of the load.

After the large end had failed, the remainder of the axle was placed in the static torsion machine and twisted to destruction. Unfortunately it was necessary to make this twist in the opposite direction from that produced in the torsion fatigue test. Fig. 5 shows the resulting rupture. The light portion is the final rupture produced

in the static torsion machine, while the dark surface is part of the conical plane produced by fatigue. Radial cracks extending inward from the roots of some of the splines will be noticed. The failure was started at a stress below the elastic limit of the axle.

Tests on Large Impact Torsion Machine—Through the courtesy of the Republic Steel Corp., a new series of tests have just been started using their large 50,000 foot-pound torsion impact machine.

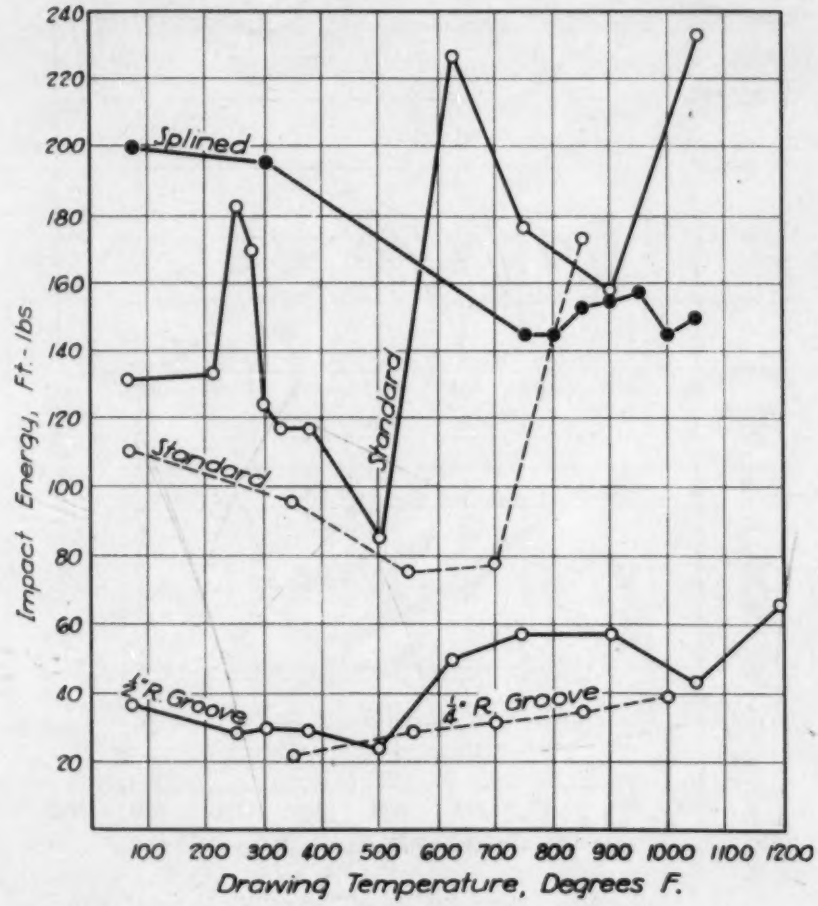

Fig. 14—Torsion Impact Energy—Small Specimens.

In these tests regular axles of different steels, heat treated for different degrees of hardness, will be broken by a single blow in torsion impact.

The machine used is similar to the original Carpenter impact machine, except that it is larger. The energy consumed in breaking the specimen is computed from the reduction of the speed of the flywheel. The data so far obtained are insufficient to warrant much discussion.

A few points, however, may be mentioned briefly. Figs. 6, 7, and 8 show the fractures of the first three axles tested. It will be noted that in spite of the high velocity at which the specimens were broken that each axle took a considerable twist. The amount of the twist and the size of the rough fracture in the center varies with

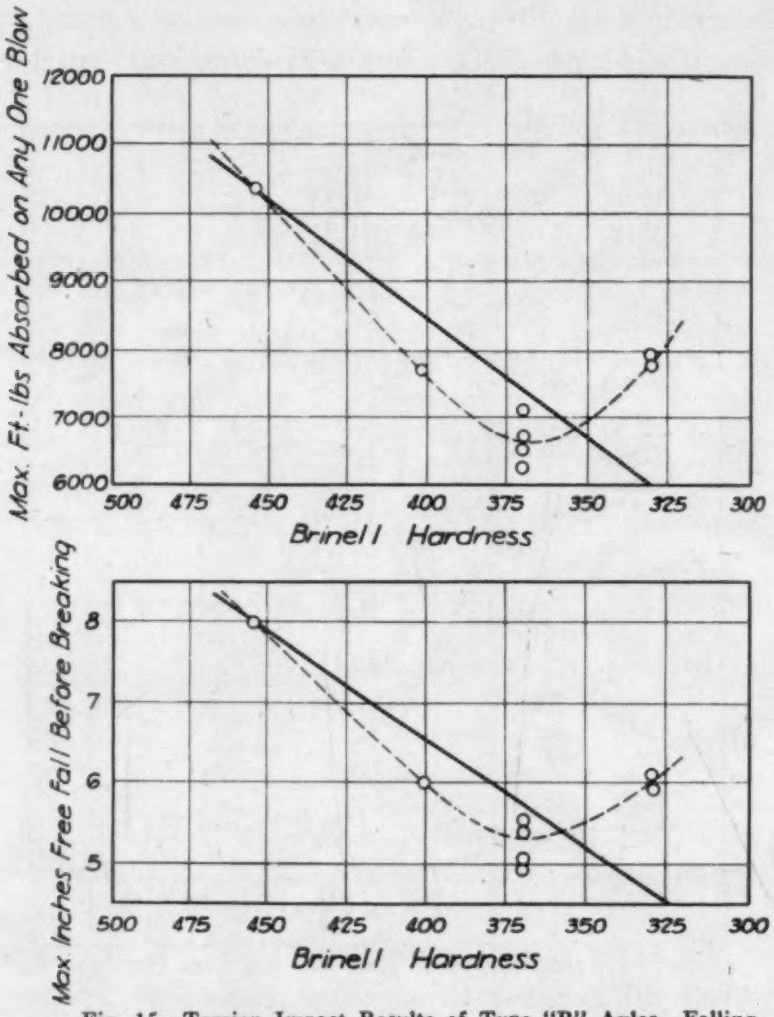

Fig. 15—Torsion Impact Results of Type "B" Axles—Falling Weight Method.

the type of steel and the hardness of the axle. These fractures do not show tears from the roots of the spline, such as were found in the axle broken in torsion fatigue.

While there are not sufficient data available at the present time to draw any definite conclusions, it might be remarked that of the three axles mentioned the one with the softest core showed the highest impact strength.

It is too early to predict what these tests will show. It is

hoped, however, that by this method of testing it will be possible to determine whether the velocity of application of stress has any effect upon the practical strength of the axles. The writer is inclined to believe that the composition and hardness of the steel, which produced the highest figures on the static torsion test, will not necessarily show up the best on the torsion impact test.

In comparing the optimum value for hardness as determined by static tensile and tensile impact tests, the velocity of application of the load did not seem to be important. It should be remembered

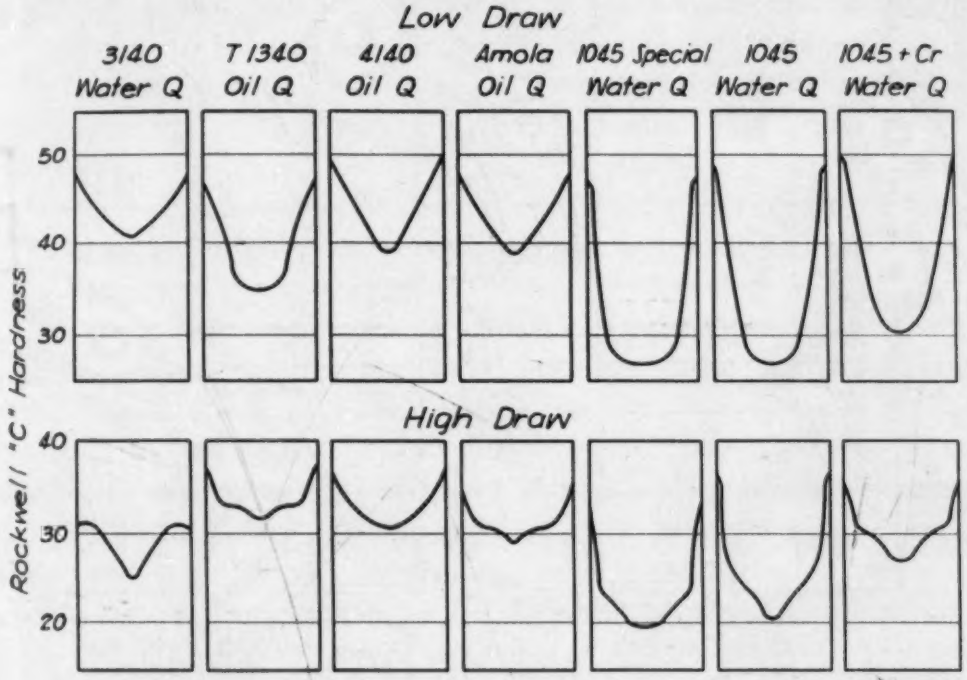

Fig. 16—Penetration of Hardness on Type "C" Axles. (Rockwell Hardness Readings Taken Every ⅛ inch from Surface to Center).

that in both cases the stresses are quite uniformly distributed. Under such conditions there is no need for great plasticity. In the case of axles, however, there is a concentration of stress at the roots of the splines.

Assuming that the function of plasticity is to permit some yielding of the more highly stressed fibres, it seems quite probable that the time factor may be important. When the load is applied at a high velocity, the higher stressed fibres may not have time to yield sufficiently unless the material is very plastic. While this point has not been definitely proven by the present investigation, it is hoped that further data will either confirm or disprove this.

While recommending this type of test as a possible approach

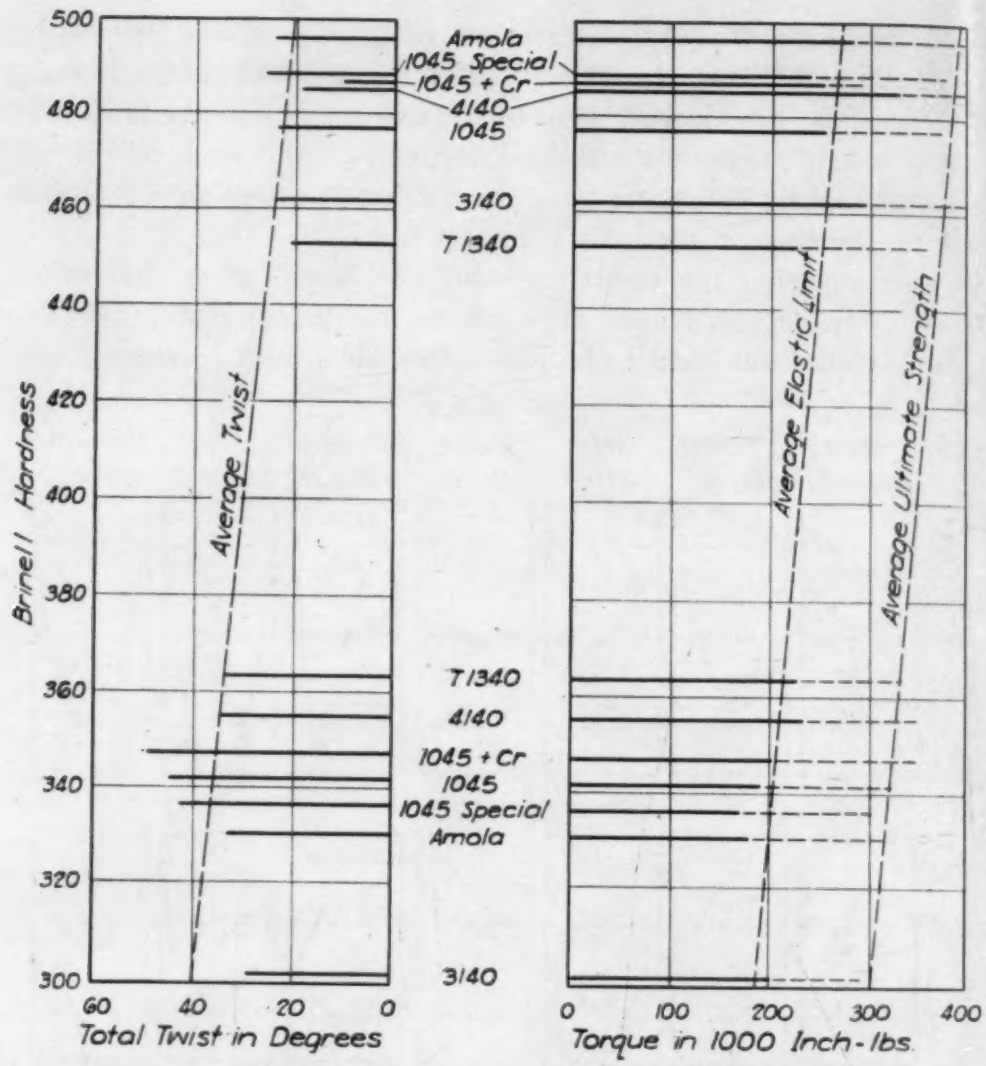

Fig. 17—Static Torsion Results of Type "C" Axles. (Note:—In the Right Hand Graph Solid Lines Represent the Elastic Limits; Solid Plus Dotted Lines Represent Ultimate Strength).

to axle problems, it may be well to give a word of caution with regard to testing full size axles. With small specimens it is permissible to allow the broken end of the specimen to fly against the sheet metal guard. With full size axles additional safeguards must be provided.

Design of Axles

While it is the primary duty of the engineering metallurgist to select steels, prescribe heat treatments, and specify physical properties, the question of design cannot be overlooked. It must be remembered that the design should be as economical as possible with-

out decreasing the practical life of the axle. In one point the design of tractor axles may differ from that of axles for trucks and pleasure cars. The weight of the axle is not so important in the case of tractors, but the total cost should be kept down.

It is obvious that there must be localized stresses at the roots of the splines. Truck and pleasure car axles are usually designed so that there is a neck near the spline end, which is smaller than the pitch diameter and sometimes smaller than the root diameter of the spline portion. Such a design may cause the failures to occur in the neck rather than in the spline. It has been found, however, that when some of the tractor axles are designed this way, the strength as determined by the static torsion test, is somewhat lower. It may be, however, that such an axle would have a higher resistance to shock and fatigue, although this point has not been definitely established. It must also be remembered that producing a neck in the axle, either by forging or machining, may mean an increase in the cost of the product.

In this paper considerable attention has been given to the merit of plasticity in permitting the yielding of the more highly stressed fibres. Plastic flow, however, can only occur when these highly stressed fibres are loaded beyond their elastic limit. It may be more desirable to design the axle so that a similar amount of deformation can take place below the elastic limit. In other words, a long slender axle may actually be stronger in service than a shorter one of a larger diameter, in spite of the apparent higher strength of the latter.

Conclusions

While this investigation has been concerned merely with tractor axles, it is believed that some of the fundamentals involved apply to many other parts as well. The following are listed as some of the conclusions from this investigation.

(1). There is no mathematical formula which can be used to determine the best type of steel or the best combination of physical properties for all parts. The selection of the steel and the physical properties depends upon the design of the part and the type of service to be encountered.

(2). When stresses are uniformly distributed, the highest hardness is associated with the greatest working strength. This applies both to surface hardness and penetration of hardness.

(3). The function of plasticity is to permit sufficient deformation of highly stressed fibres to permit a more favorable distribution of the stress.

(4). The optimum amount of plasticity depends upon the concentration of stress, or the severity of the notch effect. When the latter is very severe the maximum plasticity is required. This may mean a low surface hardness, or a shallow penetration of hardness, or both.

(5). The effect of velocity of application of the load is not proven by the tests so far conducted. It is suggested that high velocity of application may increase concentration of stress where the design is such that some concentration of stress already exists.

(6.) Even limiting the problem to axles, it is found that the selection of steel and physical properties depends upon the design of the axle. For an axle such as shown in Fig. 1-A made of S.A.E. 3140 steel, a surface hardness of 400-450 Brinell seems to be the best. For an axle such as shown in Fig. 1-C made of the same steel, a lower hardness is probably better as the harder axles shattered.

(7). For any given article the ideal surface hardness depends not only upon the design and type of service, but also upon the characteristics of the steel employed. The best hardness for one type of steel may not be the best for other types.

(8). In general steels which give a deep penetration of hardness, produce axles with higher static torsional strength but lower plasticity. There are exceptions to this rule. Certain steels are more brittle than others in certain ranges of hardness.

Acknowledgment

The writer wishes to acknowledge his indebtedness to the following men for their assistance and advice:—Messrs. G. V. Luerssen and O. V. Greene of The Carpenter Steel Co. for small torsion impact tests; Messrs. N. L. Deuble, R. Sergeson, and R. A. Carr of The Republic Steel Corp. for torsion impact tests on axles; Messrs. A. E. Dentler and Wm. Trongeau of The W. H. Miner Co. for the falling weight torsion impact tests; A. S. Jameson of West Pullman Works, International Harvester Co. for Izod tests; R. E. McGee of Farmall Works, International Harvester Co. for repeated impact tests; and the entire staff of the writer's laboratory, particu-

larly Messrs. Frank Sailer, E. H. Snyder, and R. M. Murphy, who assisted in planning and conducting tests.

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DISCUSSION

Written Discussion: By Gordon T. Williams, metallurgist, Cleveland Tractor Co., Cleveland.

Mr. Knowlton is to be congratulated on this excellent report of his painstaking and detailed investigation. Only by such studies will we ever be able to correlate the results of standard physical tests on specimens with each other and with service reports; actual field service has a disconcerting disregard for small specimen tests. Particularly is this true with tractors.

It is surprising to see no report of "starry" torsional fatigue failures through splined sections; this type of fracture predominates in the writer's experience.

In Fig. 16 of the paper showing hardness penetration, S.A.E. 1045 has such a shallow case that we would fear fatigue failures originating just below the surface, as occasionally with carburized pieces. Certainly such a material would be dangerous when splines are cut after heat treatment; root of spline stress concentration would fall in the weaker core material.

A factor of prime importance in the heat treatment of axle shafts is distortion; has the author made any observations of relative distortion of the

various steels? We would appreciate his comments regarding comparative effect of alloying elements and varying depth of hardening on distortion.

The axles failed in service by fracture through the splined section; would the stress concentration which caused this be relieved by designing with a smaller, long center section, permitting deflection and equalization of stresses? The author has indicated that he may test designs of this sort; the results will be very enlightening.

Written Discussion: By W. H. Clapp, California Institute of Technology, Pasadena, Calif.

Mr. Knowlton has conducted a very thorough and comprehensive series of tests and has contributed some interesting and valuable information. He is to be congratulated also for reviewing in his paper many practical considerations with regard to failure with which both the metallurgist and the designer should be more familiar.

The writer wishes to raise a question, applicable not only to Mr. Knowlton's paper but to nearly all reports covering the use of steel for a particular service. Is anything gained by making routine mechanical tests upon standard steel specimens? These data for tension, torsion, fatigue and notched bar impact are well known. The problem is not to determine the behavior of a few pieces of steel or even of different heats, but to determine the long range performance of different steels for a particular service.

Usually the conditions of loading and stressing are very different from those of the standard tests. In this case the axles are rather intricate machined forgings having several diameters and end splines. They are not uniformly hard either along the bar or on the cross sections. The loading is torsional with occasional shock overloads. High stress concentrations are set up at the bottom of the splines and the stress concentration factor is problematical. These stresses are superimposed upon the internal stresses in the material as a consequence of its heat treatment. For such a complicated situation how is it possible to use the results of routine testing as measures of performance in service?

Of course, if one steel showed uniformly higher mechanical properties, such as yield point, elongation, endurance limit and impact strength, and if that steel machined easily and had good depth hardening properties it would, at least, be a strong contender. Actually, this is never the case. If one will tabulate these various properties for the different machinery steels and then compare these steels on the basis of some similar property, such as tensile strength, he will be amazed to see how little variation exists in the other properties for the different steels. For a given range in drawing temperature some steels will show a much wider range of tensile strength than others, but at a given tensile strength none of these steels will vary appreciably in yield point, elongation, endurance limit, etc. In fact the variation between the steels for any one property is much less than the departure from the norm which might be expected in any one of them. Now can we say that because one of these steels shows a slightly higher elongation and lower yield point or a higher endurance limit and lesser notched bar strength, it is therefore superior for some such service as tractor axles?

The writer has long felt that it would be a boon to the designer to have

more information about the dependability of steels. If a steel is heat treated to give an anticipated yield point strength of 120,000 pounds per square inch how great a departure from this value may occur? Are some steels characterized by a wide spread and others by a much narrower one? Certainly enough mechanical testing has been done to furnish us with coefficients of dependability for the machinery steels.

Written Discussion: By H. W. McQuaid, Republic Steel Corporation, Massillon, Ohio.

It is always interesting to review papers in which a collection of actual test results are given. To those of us who have attempted to obtain clear cut information on materials for axle shafts, the great amount of work necessary to produce the data given is evident. We are fortunate, indeed, that this type of data can be made available for our study.

My experience in trying to determine the best analysis for a given axle application indicates that the conclusions drawn by Mr. Knowlton are correct. When the load is evenly distributed and the stresses uniform, plasticity plays little part and the shafts can be made to a higher Brinell. Tests (in buses for a large municipal application) indicated that S.A.E. 3240 shafts, having a Brinell hardness of approximately 514, gave better service than when the hardness was approximately 401. When considering the high hardness, however, we must not overlook the importance of internal stresses due to over quenching. If the high hardness is to be obtained, we should reduce the alloy and carbon content to the point where there are no initial internal stresses due to a martensitic-austenitic zone at the surface of the shaft. A depth hardness curve on quenching should show the maximum hardness at the surface. One of the toughest applications in my experience was for a certain dump truck where no high carbon fully hardened shaft could be made to perform satisfactorily. This difficulty was overcome by the use of a deeply carburized and double treated S.A.E. 3315 shaft.

In my opinion internal stresses play almost as important a part in the success of a heavily loaded axle as design. Not only are quenching stresses important, but cold work stresses due to cold straightening are probably of equal importance. For the insurance of maximum life, cold straightening should be eliminated.

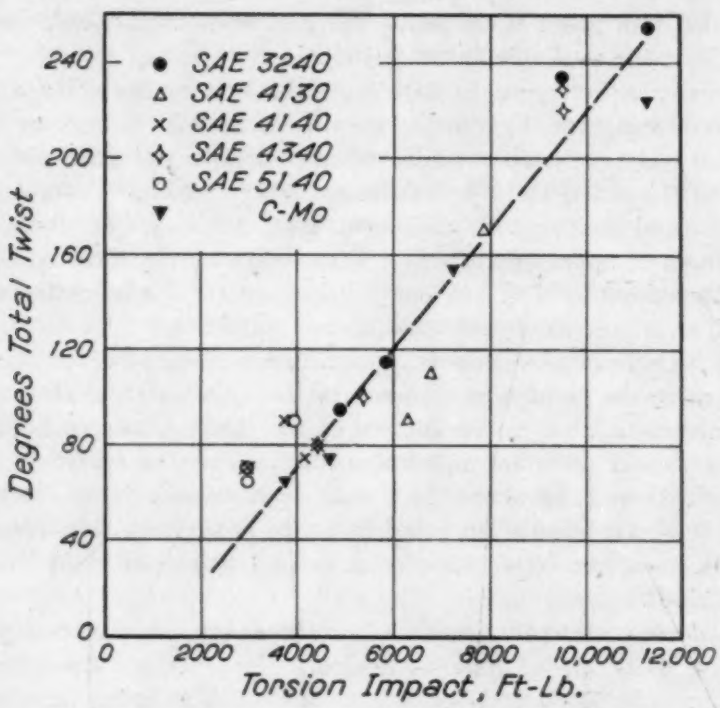
It is surprising how good the water-hardened S.A.E. 1045 analysis can be made to perform if the steel is of the proper type and the heat treating correctly performed. Millions of axle shafts of this type have given good service in competition with the low grades of alloy and in one very important truck application actually out performed the alloy type. The use of the special carbon steels is, of course, limited to smaller sections because of the effect of mass.

Mr. Knowlton's suggestion that it may be more desirable to permit the necessary deflection to take place within the elastic limit rather than by plastic deformation is very sound, and is a good argument for full floating axles. I have actually known of many cases where improved service was obtained by cutting down the shaft to provide a sufficiently long straight section of the minimum diameter to permit of enough "wind up" to take care of suddenly applied fly wheel loads.

Written Discussion: By O. W. McMullan, Youngstown Sheet and Tube Co., Indiana Harbor, Ind.

In regard to conclusion No. 8 the writer has frequently found that deeper hardening steels such as S.A.E. 3240 and 4340 will give a greater degree of twist in a static torsion test than will shallow hardening steels.

The author mentions results obtained with Republic Steel Corporation's single blow torsion impact machine. Some time ago a number of tests were made for the writer on the same machine. The various steels were tested as specimens heat treated to represent the upper and lower limits of hardness



commonly specified for commercial vehicle shafts. The specimens were made with a reduced section 3 inches long and one inch in diameter and were heat treated after being machined to size. Results on some of the steels tested are shown in the figure in which foot-pounds of energy are plotted against degrees of twist in the 3-inch gage length.

It will be noticed that the points fall fairly close to a straight line regardless of material, treatment, or hardness. From the hardness figures given in the table it will be noticed that energy required increases with decrease in hardness. Center hardness values are not available but were probably close to those obtained on the surface. It is not known if specimens of lower hardness would continue to follow along the line drawn in the figure. Neither is it known if the materials, similarly treated, would occupy the same relative positions as regards degree of twist if tested in static torsion. I would like to ask the author if he has any evidence bearing on these points.

Steel	Specimen	Brinell	Torsion Impact Ft.-Lbs.	Total Twist
S.A.E. 3240	A	444	4875	95
	B	444	5820	115
	C	363	11280	255
	D	363	9500	235
S.A.E. 4130*	A	415	6289	90
	B	415	6755	110
	C	363	7864	170
S.A.E. 4140	A	444	3728	90
	B	444	2955	65
	C	363	4399	80
	D	363	4112	75
S.A.E. 4340	A	444	4112	75
	B	444	5349	100
	C	363	9500	220
	D	363	9500	230
S.A.E. 5140	A	415	2955	70
	B	415	3920	90
	C	363	2955	65
C-Mo	A	444	3728	65
	B	444	4685	75
	C	363	7213	155
	D	363	11280	225

*Water-quenched; others oil-quenched.

Oral Discussion

H. F. MOORE:[1] Many experimenters have tried to run rotating-beam fatigue tests on specimens not reduced at the center of length, and have found fracture to take place at unduly low computed stress at or near the edge of a collet jaw holding the specimen. At that jaw-edge there are highly localized stresses which consist of an intense compressive stress over a minute area. As the specimen rotates, this minute area of intense compression moves round the specimen. Now a small distance below the surface subjected to this high localized compressive stress there are set up shearing stresses on diagonal planes. These shearing stresses can be estimated by the Belajef-Hoersch extension of the Herz formulas for intensified pressure under localized compression.[2]

Now when cold working of the surface takes place the stress distribution is modified from what is the distribution under elastic conditions, and, in general, the localized stresses are reduced by the plastic action accompanying cold work. This is illustrated by the effect of cold rolling of railroad rails under traffic. The localized shearing stresses under wheel loads are reduced by the plastic action of cold rolling, and the tendency to start fissures close to the wearing surface of the rail is diminished, so that when and if internal fissures do start, they start at minute cracks much further below the wearing surface of the rail than is the zone of maximum shearing stress under a concentrated load. Cold work properly done seems decidedly effective in lessening the probability of the start of a spreading crack below a surface subjected to high localized compression.

Author's Reply to Discussion

I wish to express my appreciation of the criticisms given by the various discussers. Taking these up in order, I would like to add the following comments:

[1]Professor of Engineering Materials, University of Illinois, Urbana, Illinois.

[2]See Bulletin 212 of the University of Illinois Engineering Experiment Station.

I do not know why we have not encountered the "starry" fractures in service or fatigue failures of tractor axles. This type of failure seems to be more common in the case of our truck axles but we do not know why.

Mr. Williams' comments with reference to the shallow penetration of hardness in the S.A.E. 1045 steel axles is worthy of thoughtful consideration. While we have favored alloy steel for axles, it must be recognized that plain carbon steel has been successfully employed. With regard to relative distortion of the different steels we have no data to submit. In this investigation all of the axles, which were heat treated for the higher hardness range, were machined completely except for grinding, before heat treating and no trouble was experienced with distortion of the splines. All of the axles fit the splined chucks used in the torsion tests.

With regard to changing the design so as to have a long center section smaller in diameter than the splined section, it may be said that this would probably produce greater resistance to shock. On the other hand, experiments which have been conducted since this paper was written, have indicated that axles with the reduced section may actually be weaker in fatigue than the straight axles. If the stresses applied in the fatigue test are the same for axles of the two designs, the degrees of twist will be greater for the axle with the reduced section. It is suggested as possible that the greater angular movement between the splined and the small sections may itself be responsible for lowering the fatigue strength.

Mr. Clapp brought up the question of the value of routine laboratory tests. We are willing to grant that the ideal test is actual service over a long period of time. Such tests are long and expensive and very difficult to standardize. It is desirable, therefore, to devise laboratory tests which will in some measure at least predict the performance in service. Such was the objective of the investigation described in this paper. We believe that the torsion fatigue test does simulate service sufficiently to be used as a criterion for judging the merits of different steels. Some of the other tests did not seem to correlate with service, while others did give important information.

We are also willing to grant that tensile strength, yield point, and hardness are proportional to each other regardless of the types of steel. We have found some variation in the degree of plasticity at the same hardness. Still more important, there is a great difference in the penetration of hardness in different steels as was shown in Fig. 16.

With regard to the "dependability" of different steels, we have no data to offer. It would require actual use of different steels in production over a period of considerable time before such data could be acquired.

With regard to Mr. McMullan's question on torsion impact strength, we have no further data to submit. As stated in the paper we believe that the optimum hardness is a function of the design of the specimen. It seems entirely possible that the 1-inch diameter 3-inch long specimen would show a lower optimum hardness than a ¼-inch diameter 1-inch long specimen.

We wish to express our appreciation of the discussions given by Professor Moore, and Mr. McQuaid, but have no comments to offer on their discussions.

WHICH GRAIN SIZE?

By C. H. Herty, Jr., D. L. McBride and E. H. Hollenback

Abstract

The hardenability of a steel during heat treatment is profoundly affected by the grain size of the steel. A common method of judging grain size is the McQuaid-Ehn test, in which the steel is carburized at 1700 degrees Fahr. for 8 hours.

This paper points out that the size of grain which determines hardenability is that present in the steel at the moment of quenching, and not that determined under conditions quite different from those encountered in normal heat treating operations.

It is experimentally shown that for a given analysis hardness penetration is a direct function of the grain surface area per unit volume of steel at the quenching temperature. The surface area is a function of the grain size of the steel. Surface hardness is shown to be essentially independent of grain size, provided the critical cooling rate is exceeded. This rate is higher for fine-grained than for coarse-grained steels. If the cooling rate is decreased below the critical value, due to gas or scale interference with heat transfer, soft spots will occur.

SINCE McQuaid and Ehn's early work on carburizing steel, the interest in and application of the McQuaid-Ehn test has increased to the point where this test is specified on a large tonnage of various grades of steel, especially the carburizing and forging grades. In many other grades the test is used to determine the grain size of a sample, and from this grain size certain general characteristics of the steel are predicted.

In using this test the inference is frequently drawn that the observed grain size is that which exists in the steel during heat treatment. Variations in response to heat treatment for a given McQuaid-Ehn grain size are common, due to the fact that the standard McQuaid-Ehn test does not indicate the grain size existing in the steel except under the specific conditions of the test. For exam-

Of the authors, C. H. Herty, Jr., and D. L. McBride are associated with Development and Research Department, Bethlehem Steel Co., Bethlehem, Pa., and E. H. Hollenback is Assistant Engineer of Tests, Cambria Plant, Bethlehem Steel Co., Johnstown, Pa. Manuscript received January 25, 1937.

ple, steels showing mixed grain in the McQuaid-Ehn test have been condemned for quenching operations because it was believed that the mixed grain structure would cause cracking or excessive warpage. These steels are in general uniformly fine-grained at the usual heat treating temperatures, 1600 degrees Fahr. (870 degrees Cent.) or lower, and behave as such in heat treating operations.

The purpose of this paper is to show the relationship between hardenability and grain size and to emphasize that the grain size important to hardenability is that formed at the heat treating temperature, and not that shown by the McQuaid-Ehn test, unless the steel is to be heat treated under the conditions of that test.

The McQuaid-Ehn Test

The McQuaid-Ehn test was first intended to classify steels with respect to their hardening characteristics after carburizing. The test consists of carburizing a sample at 1700 degrees Fahr. (925 degrees Cent.) for eight hours, followed by a box cool and subsequent examination of the carburized case and the core of the test specimen. The grain size is usually judged by the case, although the core and the transition zone are sometimes used. During carburizing the steel absorbs carbon to the extent that the case, usually about 0.03 inches in depth, contains from 1.00 to 1.10 per cent carbon. The core has the carbon content of the sample used and the transition zone decreases in carbon content from case to core between these two limits. With the amount of carbon present and with the slow cooling in the box cementite is precipitated at the austenitic grain boundaries and the observed grain size is, therefore, the true austenitic grain size of the carburized case at the temperature of carburizing. However, the case has increased in carbon content from that of the sample to 1.00-1.10 per cent, whereas the core has the carbon content of the sample. This change in carbon content has in many instances an effect on the grain size of the steel at the carburizing temperature. Furthermore, it must be remembered that the observed grain size is established during eight hours at temperature.

When the carburizing temperature is varied, the observed grain size also varies, i.e., the lower the carburizing temperature the finer the grain. As will be shown presently this change in grain size with temperature is of very great importance in heat treating operations.

The microstructure of the grain after the carburizing treatment is often used to judge the character of the steel. The familiar terms

"normal" and "abnormal" are used to differentiate between two extremes in structure, as follows: A "normal" structure is one in which the grains are clearly outlined with a continuous network of cementite of fairly uniform width, the internal grain structure exhibiting no areas of divorced ferrite. When the pearlite is lamellar the alternate plates of carbide and ferrite are in straight parallel lines. An "abnormal" structure is one in which divorced ferrite and globular cementite appear in either or both network and internal grain structure, the network usually being of variable width and discontinuous. Inasmuch as the carbon is in almost all instances, uniformly dissolved in the austenitic grain after an eight-hour soak at 1700 degrees Fahr. (925 degrees Cent.), it is evident that the McQuaid-Ehn test will show the relative rates of coalescence of precipitated carbides during the cooling period. A normal steel has a very slow rate of coalescence of carbide, an abnormal steel a rapid rate. It, therefore, follows that the degree of abnormality can be controlled by varying the cooling rate after carburizing, and such is actually the case.

The fallacy of indiscriminately utilizing the grain size observed in the McQuaid-Ehn test for predicting properties developed in treatments other than carburizing will now be demonstrated by a consideration of the austenitic grain size characteristics of steel, and the relation between hardenability and austenitic grain size.

Effect of Temperature on Grain Size

As indicated above, the temperature at which the carburizing test is run determines the resulting grain size for a given sample. The austenitic grain size may be estimated by a number of simple test methods as described by Bain and Vilella.[1] In the following experiments a simple normalizing treatment has been used. Fig. 1 shows the change in grain size with temperature in the carburizing test for four types of steel. The steels were chosen so that one was coarse-grained (A); one mixed-grained with a tendency to be coarse (B); one mixed-grained with a tendency to be fine (C); and one fine-grained (D), *in the standard McQuaid-Ehn test.* The analyses of these steels are as follows:

Steel No.	Carbon	Manganese	Phosphorus	Sulphur	Silicon	Lbs. Alum Per Ton
A	0.51	0.46	0.018	0.029	0.16	0.55
B	0.48	0.82	0.020	0.028	0.22	0.58
C	0.42	0.78	0.019	0.037	0.16	1.40
D	0.39	0.62	0.014	0.028	0.17	1.64

[1]Metals Handbook, 1936, page 584.

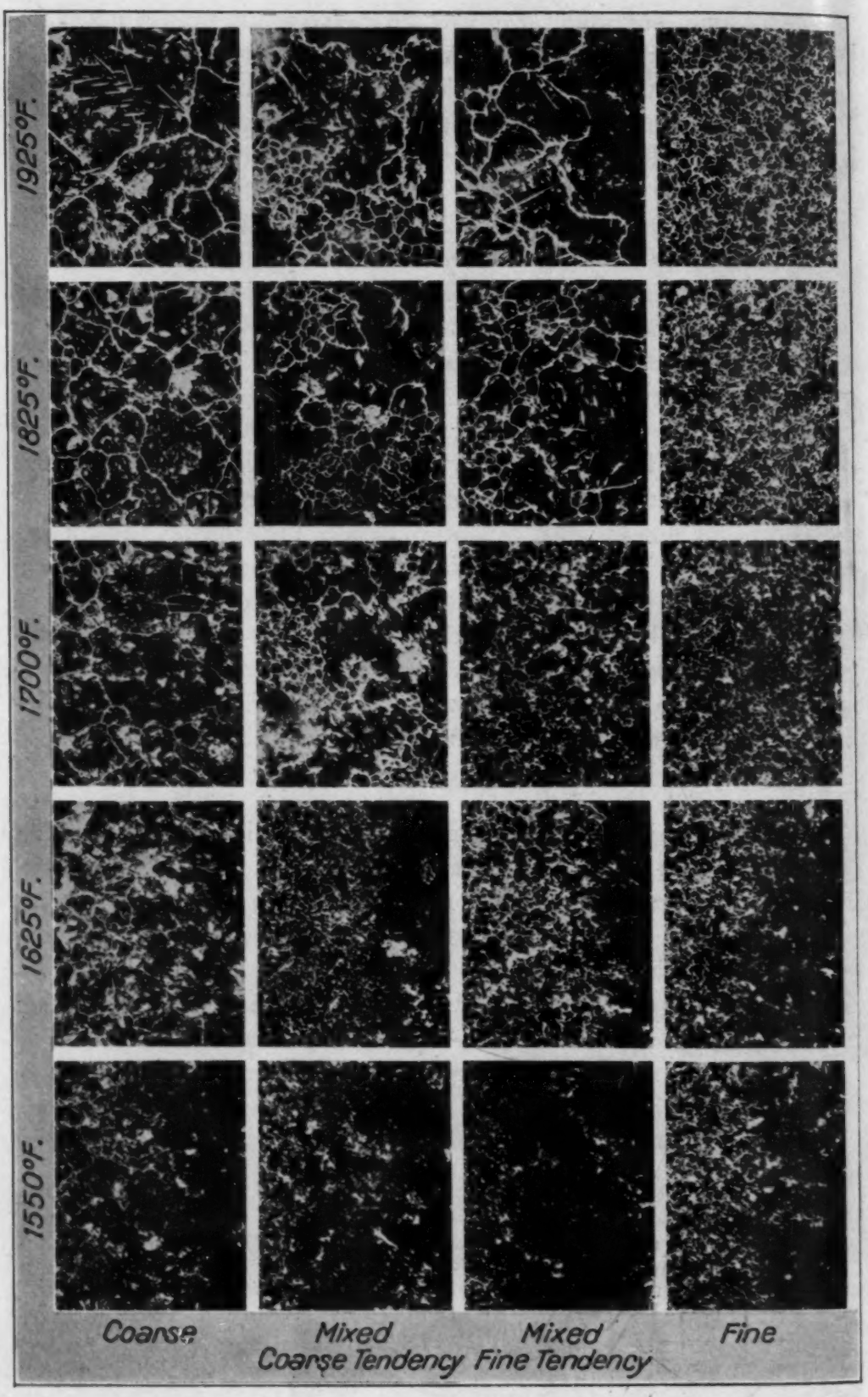


Fig. 1—Effect of Temperature on Grain Size of Carburized Steels. × 100. Reduced 43 Per Cent in Reproducing for Printing.

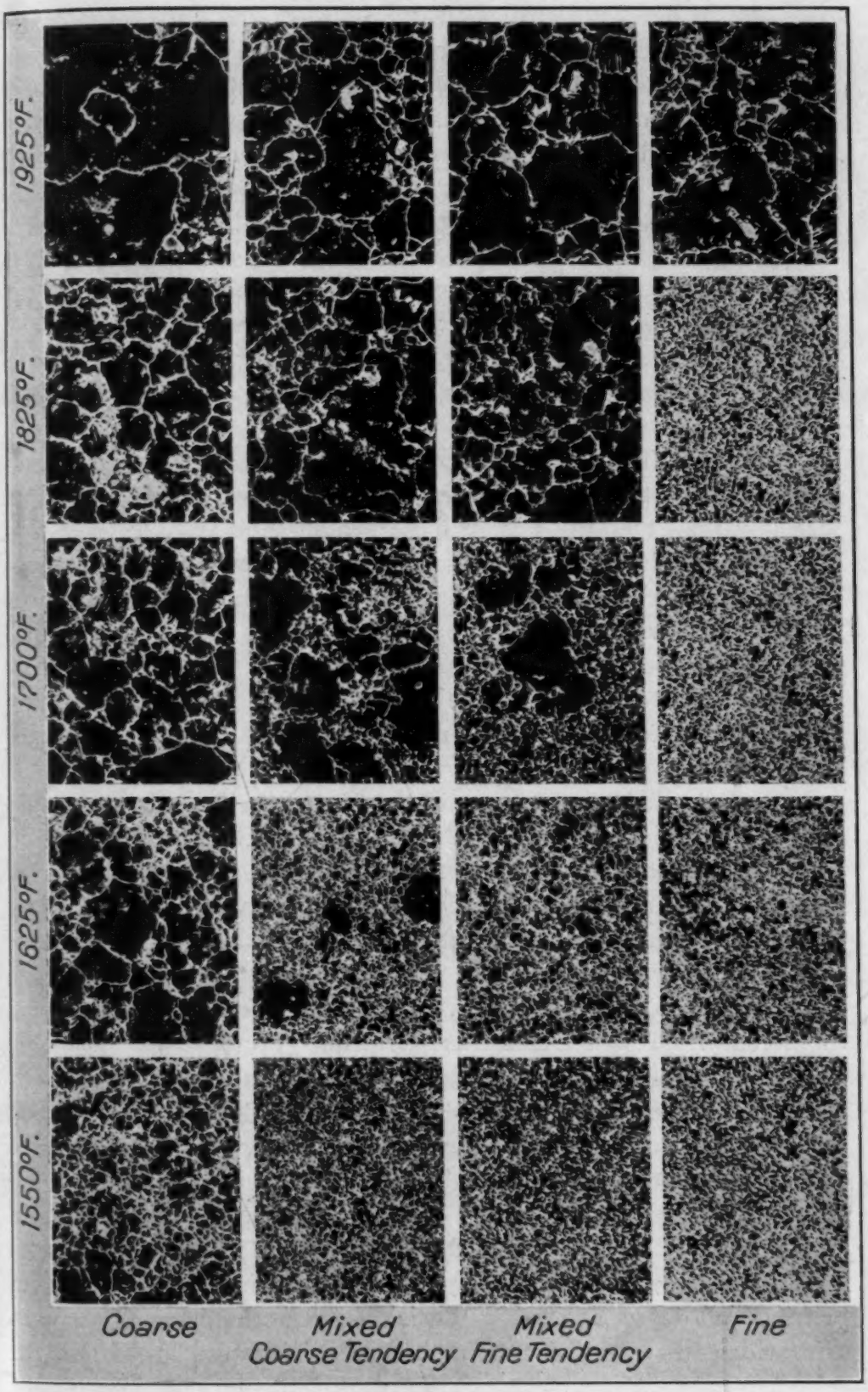


Fig. 2—Effect of Temperature on Grain Size of Normalized Steels. × 100. Reduced 43 Per Cent in Reproducing for Printing.

Fig. 2 shows the change in grain size of the same steels in the normalized condition. In this treatment the steels were held at temperature for one hour. The close parallelism between the two types of heat treatment is readily seen.

Steel A. This steel is coarse-grained in the McQuaid-Ehn test and is medium-grained, and rather uniformly so, when carburized at 1550 degrees Fahr. (845 degrees Cent.). The most pronounced coarsening takes place between 1550 and 1625 degrees Fahr. (845 and 885 degrees Cent.). In the normalized condition, the steel appears quite similar to the carburized condition, except that the sample normalized at 1550 degrees Fahr. (845 degrees Cent.) is somewhat finer grained than that carburized at 1550 degrees Fahr. (845 degrees Cent.).

Steel B. This steel is mixed grained with a marked tendency to be coarse in the McQuaid-Ehn test. As the carburizing temperature is raised, the steel becomes coarser but fine areas remain. At lower carburizing temperatures the steel is fine-grained. In the normalized samples much the same condition holds as in the carburized tests, with two exceptions. First, there are no fine-grained areas at the two higher normalizing temperatures and second, there are coarse spots in the sample normalized at 1625 degrees Fahr. (885 degrees Cent.) which do not appear in the sample carburized at 1625 degrees Fahr. (885 degrees Cent.).

Steel C. This steel has a mixed grain with a fine grain tendency in the McQuaid-Ehn test. At higher temperatures the steel is coarse-grained and at lower temperatures, fine-grained. The normalized structures show a perfect parallelism with the carburized structures for this steel, and at 1550 degrees Fahr. (845 degrees Cent.) the normalized structure shows that the steel is typically fine-grained. At 1625 degrees Fahr. (885 degrees Cent.) there is a very slight coarsening, but the steel is still a fine-grained steel.

Steel D. This steel is a fine-grained steel in the McQuaid-Ehn test and just begins to coarsen at about 1925 degrees Fahr. (1050 degrees Cent.) under carburizing test conditions. In the normalized condition the steel is fine-grained at the four lower normalizing temperatures, except for a spot of coarse grain at 1825 degrees Fahr. (995 degrees Cent.), but at 1925 degrees Fahr. (1050 degrees Cent.) the steel is definitely coarse-grained. In comparing the 1925 degrees Fahr. (1050 degrees Cent.) carburizing and normalizing treatments, it is evident that some factor in the carburizing test has caused the

coarsening temperature to be raised somewhat over that of the normalized steel.

From these results it is evident that steels which show mixed grain in the McQuaid-Ehn test will be fine-grained when heat treated from 1550 degrees Fahr. (845 degrees Cent.) or lower, particularly if the mixed grain has a tendency to be toward the fine-grained side. If the mixed grain in the McQuaid-Ehn test is preponderantly coarse, a uniform grain size is obtained at 1550 degrees Fahr. (845 degrees Cent.), slightly larger than that of steels which are finer grained in the McQuaid-Ehn test. In a steel of this type there will probably be spots of coarse-grained material in a matrix of fine grains when heat treated above 1550 degrees Fahr. (845 degrees Cent.).

In both tests it is the austenitic grain which is outlined, in the carburized pieces by cementite, in the normalized pieces by ferrite. Tables I and II give the grains per square inch for all samples. These grain sizes were either counted or estimated from comparison with samples already counted.

Table I
Grain Size of Samples Carburized at Various Temperatures

Steel No.	1550 Degrees Fahr.	1625 Degrees Fahr.	1700 Degrees Fahr.	1825 Degrees Fahr.	1925 Degrees Fahr.
(A)	30.0	8.1	6.7	2.9	2.6
(B)	*200	*180	18.0	5.5	4.4
(C)	*200	182	*100	4.2	1.9
(D)	*200	196	*150	*85.0	67.0

*Estimated.

Steels which are coarse-grained in the McQuaid-Ehn test may show widely different grain sizes at heat treating temperatures, as recorded in Table III.

It is apparent from these data that the results of the McQuaid-Ehn test on steels of Type A may be entirely misleading. For ex-

Table II
Grain Size of Normalized Samples, Heated for One Hour at Various Temperatures and Air-Cooled

Steel No.	1550 Degrees Fahr.	1625 Degrees Fahr.	1700 Degrees Fahr.	1825 Degrees Fahr.	1925 Degrees Fahr.
(A)	88.0	13.2	8.3	5.5	3.4
(B)	165	*150	*80	5.5	3.3
(C)	*200	*170	*140	*8.5	3.4
(D)	171	*170	*170	135	2.5

*Estimated.

Table III

Change in Grain Size of Coarse-Grained Steel with Temperature

Steel No.	Analysis C	Mn	P	S	Si	McQuaid-Ehn Grain Size	Grains Per Square Inch at 100 × When Normalized at 1550 Degrees Fahr. (845 Degrees Cent.)
J	0.39	0.73	0.016	0.031	0.07	2	15
AD	0.41	0.81	0.034	0.039	0.17	2	25
A-98	0.49	0.74	0.016	0.032	0.22	2–4	29
AO	0.51	0.77	0.021	0.031	0.24	2–4	29
X	0.33	0.44	0.011	0.029	0.16	2–4	32
B-98	0.45	0.78	0.018	0.035	0.19	2–4	48
B-100	0.45	0.72	0.019	0.030	0.20	2–4	53
B-99	0.46	0.81	0.020	0.028	0.19	2–4	70
C-3	0.51	0.46	0.018	0.029	0.16	2–4	88

ample, different responses would be expected from steels C-3 and J when heat treated at 1550 degrees Fahr. (845 degrees Cent.), due to the wide variation in grain size at this temperature. Many of the variations in heat treating characteristics of steels of any given McQuaid-Ehn test rating may be attributed to differences in grain size at the heat treating temperature which were not indicated by the McQuaid-Ehn test.

Effect of Time on Grain Size

It has been shown[2] that the major change in grain size within the coarsening temperature range occurs during the first few minutes at temperature. At lower temperatures, the rate of grain growth with time decreases markedly; and for steels which are mixed-grained or fine-grained at 1700 degrees Fahr. (925 degrees Cent.) or higher, the rate of grain growth is exceedingly slow at lower temperatures.

In heating a series of coarse and fine-grained (McQuaid-Ehn test) 0.40 per cent carbon steels to various temperatures, the actual time required at temperature to give the major amount of coarsening is shown in Table IV.

Table IV

Time at Temperature for Major Coarsening (Minutes)

Temperature Degrees Fahr.	Silicon Killed Coarse-Grained	Aluminum-Treated Mixed Grain	Aluminum-Treated Fine Grain
1450	480	.....	.....
1550	120	No growth in 2880	No growth in 2700
1742	10–20	0–20	No growth in 480
1832	10	0–15	30
1925	4	....	10

[2]Cooperative Bulletin 65, Mining and Metallurgical Advisory Board, 1934.

After these time intervals there may be some change in grain size upon prolonged heating if the temperature is within the coarsening range, i.e., a steel which has six grains per square inch when heated at 1830 degrees Fahr. (1000 degrees Cent.) for fifteen minutes may show five grains per square inch when heated for one hour at the same temperature. However, if a steel is heat treated well below the coarsening temperature range, as is the case with fine-grained and some mixed-grained steels, no coarsening will occur even if the steel is held at temperature for as much as forty-eight hours.

For normal heat treating operations, involving fine or mixed-grained McQuaid-Ehn steels, it is probable that the grain size established when the piece has come to temperature will be the grain size obtained after the prescribed soaking period. In heat treating coarse-grained steels a gradual increase in the austenitic grain size will occur with time until the stable grain size is attained. The time necessary for the attainment of this stable grain size varies with the temperature, being greater for the lower temperatures.

Grain Size and Grain Surface

The most important feature of grain size undoubtedly lies in its influence on hardenability and the various properties of the steel following hardening treatments. In order to discuss this subject properly it will be advisable to demonstrate the relationship between grain size, grain surface and the reactions which take place during the cooling of steel.

In air-cooled hypoeutectoid plain carbon steels, ferrite and pearlite are formed as a result of the decomposition of austenite. If the steel is very slowly cooled, the amount of ferrite formed may be calculated from the composition of the steel. With more rapid rates of cooling, less and less ferrite is formed because the equilibrium transformation range is passed through so quickly that a large part of the austenite remains undecomposed until a high supercool has been reached. Under these conditions austenite transforms into fine pearlite, lower in carbide and richer in ferrite, than a true pearlite of eutectoid composition.

The ratio of the ferrite formed on air cooling to ferrite formed on annealing can, therefore, be used to determine the rate of decomposition of the austenite on air cooling. If two samples of the same steel be so treated that two different grain sizes are obtained,

the effect of grain size on the rate of decomposition of austenite may be determined by the ferrite ratio mentioned above. McBride, Herty, and Mehl[3] studied this effect of grain size on the amount of ferrite formed on air cooling in various sized cubes. It was found that the ratio

Per cent Ferrite formed on normalizing ÷ Per cent Ferrite formed on Annealing

when plotted against grains per square inch in the normalized sample gave smooth curves as shown in Fig. 3, indicating that the amount of decomposition was proportional to grain size. The larger the cube

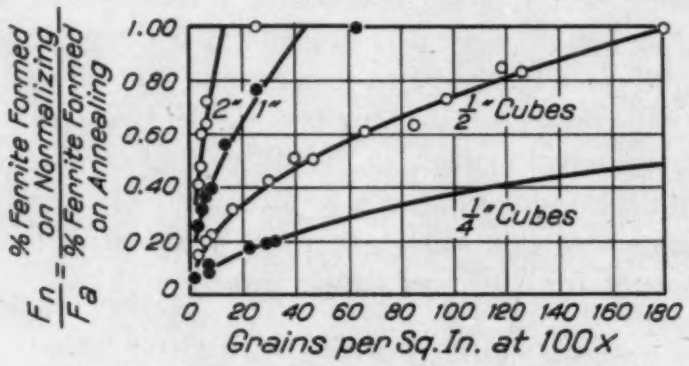


Fig. 3—Effect of Grain Size and Mass on Ratio of Ferrite Formed During Normalizing and Annealing.

the greater the amount of ferrite formed on air cooling for a given grain size. From the grains per square inch the grain surface per cubic inch of steel was calculated from the equation.

$$S = 4\sqrt{G}$$

where S = Square inches grain surface per cubic inch at × 100.
and G = Grains per square inch at × 100.

In this calculation it was assumed that the grains were spheres. The ferrite ratio when plotted against *grain surface* per cubic inch gave a straight line for each size of cube employed, as shown in Fig. 4. This was a very definite indication that the rate of decomposition of the austenite was a direct function of the grain surface per unit volume of steel.

For the readers' convenience Table V shows the relation between A.S.T.M. grain size number, grains per square inch at × 100, and square inches grain surface per cubic inch at × 100, as calculated from the above equation.

[3]D. L. McBride, C. H. Herty, Jr. and R. F. Mehl, "The Effect of Deoxidation on the Rate of Formation of Ferrite in Commercial Steels," TRANSACTIONS, American Society for Metals, Vol. 24, No. 2, p. 281.

Table V
Grain Size Relationship

A.S.T.M. Grain No.	Grains Per Sq. In. at × 100	Sq. In. Grain Surface Per Cu. In. at × 100
1	1	4.0
2	2	5.6
3	4	8.0
4	8	11.3
5	16	16.0
6	32	22.6
7	64	32.0
8	128	45.3
9	256	64.0

It should be noted that in order to double the grain surface, the important factor, it is necessary to quadruple the number of grains per square inch, i.e., an increase of two grain size numbers.

Influence of Grain Size on Hardenability

In order to obtain the maximum hardness for a given chemical composition it is necessary that the austenite decompose into martensite. In carbon steels austenite will not decompose into marten-

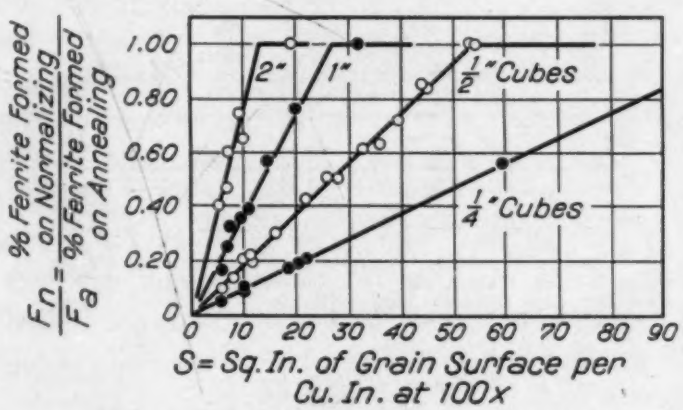


Fig. 4—Effect of Grain Surface and Mass on Ratio of Ferrite Formed During Normalizing and Annealing.

site on quenching unless the decomposition takes place at about 200 to 300 degrees Fahr. (95 to 150 degrees Cent.) or lower. This means that the steel must be cooled so rapidly that no austenite decomposition takes place *until* these temperatures are reached. The rate at which the steel must be cooled in order to fulfill the above condition is called the critical cooling rate.

It has been shown above that the rate of austenite decomposition is a direct function of the *grain surface* for any given cooling rate, the larger the grain surface, i.e., the smaller the grain size, the

more rapid the decomposition. Therefore, the critical cooling rate is greater for fine-grained than for coarse-grained steels. If the critical cooling rate at any point is reached or exceeded, grain size will have no effect upon the actual hardness values, the values then being determined entirely by the composition of the steel.

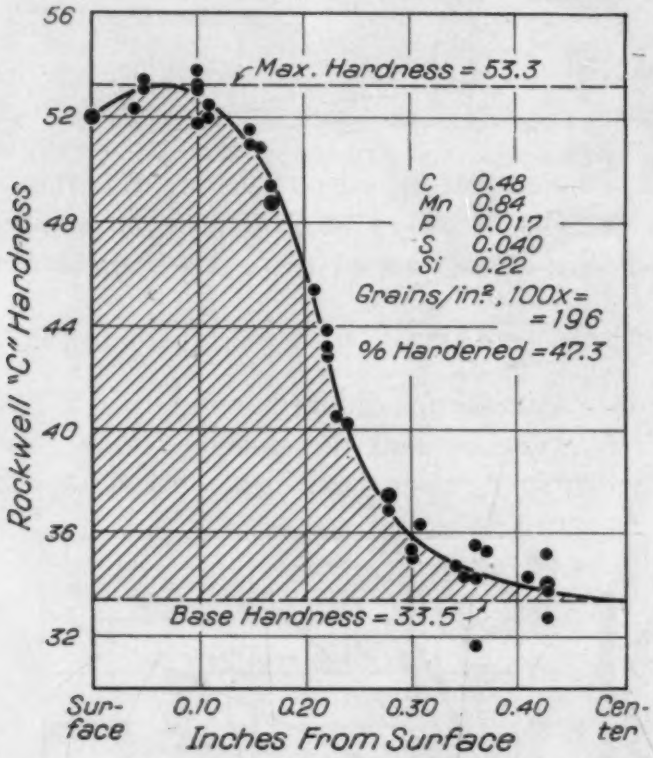


Fig. 5—Hardness Penetration Below the Surface of a 1-Inch Round Bar; 1550 Degrees Fahr. Water Quench, 500 Degrees Fahr. Temper.

Hardness Penetration

The penetration of hardness in a quenched piece may be judged by fracture and observation and by magnetic methods, or by direct hardness readings on a cross section of the hardened piece. Excellent standards can be set up for any of these tests as, for example, the Shepherd P-F ratings. In the present work, in order to obtain as accurate an evaluation as possible for hardness penetration, it was decided to use hardness readings on six radii of the cross section of a hardened 1-inch round. These hardness readings when plotted against distance from the surface give the familiar curve for hardness penetration as shown in Fig. 5. The upper horizontal dotted line of this area represents the maximum hardness found on the

hardness penetration curve. The lower horizontal dotted line, called the base hardness, is obtained for each steel from Fig. 6, which shows core and maximum hardness plotted against hardness factor. This figure was obtained by plotting the core hardness for fine-grained steels and the maximum hardness for all steels studied against a hardness factor involving steel analysis as follows:

$$3000\ C + 800\ Mn + 500\ Si + 4000\ P = \text{Hardness Factor}$$

where the elements are expressed in per cent. The reason for choosing this "base hardness" is that it is the average minimum core hardness obtainable for a 1-inch round under the heat treatment chosen, viz: a 1550 degrees Fahr. (845 degrees Cent.) water quench followed by a 500 degrees Fahr. (260 degrees Cent.) draw.

The relative hardness penetration, hereafter called the "per cent hardened," is taken as the shaded area under the hardness penetration curve, divided by the total area bounded by the heavy dotted lines on Fig. 5.

In considering the hardenability of steels which are coarse-grained through method of deoxidation rather than heat treatment, it must be remembered that the "base hardness" chosen in this work is a function of composition alone. Changes in core hardness due to grain size are automatically taken into account in calculating "per cent hardened" when the base hardness is calculated from the chemical composition for fine-grained steels.

The grain surface in which we are interested here is that which is formed at the heat treating temperature and not that shown by the McQuaid-Ehn test. This is particularly true of coarse and mixed-grained steels in that test. To determine the effect of grain size on hardness penetration a series of carbon steels were heated to different temperatures to obtain variations in austenitic grain size, were then furnace-cooled to 1500 degrees Fahr. (815 degrees Cent.), then water-quenched and drawn for one hour at 500 degrees Fahr. (260 degrees Cent.). Separate samples of each steel were heated simultaneously and were air-cooled, to determine the austenitic grain size at the time of quench. From the austenitic grain size, determined by counting, the square inches of grain surface were calculated and plotted against the per cent hardened for each sample, the per cent hardened being determined as illustrated in Fig. 5.

Fig. 7 shows the relationship between per cent hardened and square inches of grain surface. The straight line relationship be-

Table VI

Analysis of Steels Used in Hardness Penetration Experiments

Steel No.	Carbon	Manganese	Phosphorus	Sulphur	Silicon
A-98	0.49	0.74	0.016	0.032	0.22
A-99	0.38	0.75	0.027	0.031	0.20
B-13	0.41	0.78	0.020	0.034	0.23
B-66	0.37	0.70	0.026	0.029	0.21
B-72	0.38	0.78	0.018	0.040	0.21
B-87	0.57	0.77	0.017	0.032	0.23
C-13	0.37	0.74	0.015	0.030	0.20
C-15	0.44	0.72	0.017	0.033	0.19
C-23	0.46	0.73	0.035	0.035	0.20
C-29	0.58	0.71	0.024	0.033	0.17
C-31	0.39	0.71	0.018	0.034	0.20
C-37	0.61	0.76	0.017	0.028	0.21
C-38	0.57	0.74	0.014	0.030	0.24
C-61	0.45	0.73	0.014	0.024	0.19
C-68	0.45	0.71	0.020	0.032	0.19
C-86	0.59	0.79	0.020	0.030	0.21
C-91	0.55	0.54	0.017	0.035	0.21
D-12	0.44	0.68	0.015	0.030	0.20
D-13	0.48	0.76	0.019	0.036	0.23
D-15	0.38	0.75	0.017	0.033	0.19
D-23	0.48	0.84	0.017	0.040	0.22
D-30	0.35	0.68	0.034	0.038	0.18
D-32	0.46	0.74	0.022	0.026	0.20
D-33	0.44	0.74	0.015	0.035	0.18
D-47	0.58	0.74	0.018	0.032	0.22
D-50	0.43	0.71	0.016	0.040	0.26

tween per cent hardened and grain surface, over the entire range of steel analysis given in Table VI, shows definitely that hardness penetration is a function of grain surface alone.

As shown in Fig. 6, a 0.60 per cent carbon steel has a higher base hardness and maximum hardness than a 0.30 per cent carbon steel, and the base hardness and maximum hardness increase regularly with increase in percentage of hardening elements. If chemical composition within the range studied directly affected hardness penetration, the steels shown in Table VI would give points falling on different lines on Fig. 7 instead of falling on a single line as they do. In other words, a change in chemical composition, within the range of analysis studied, simply raises or lowers the level of actual hardness values in the piece, without changing the type of penetration curve.

In general, fine-grained steels are considered to be shallow hardening and coarse-grained steels deep hardening. Fig. 7 shows that the depth of hardness penetration for a given analysis of steel is directly proportional to the grain surface involved. This in turn is a function of the grain size at the heat treating temperature. Having determined this grain size, the actual per cent hardened may be easily determined by calculating the grain surface and referring directly to Fig. 7.

All this discussion refers to the range of analyses shown in Table VI. The effect of alloying elements on hardness penetration is not covered in this paper.

Surface Hardness

It was stated in the previous section that the surface hardness of a quenched steel would be independent of the grain size and dependent only on chemical composition, provided the critical cooling rate was reached or exceeded during quenching.

If the critical cooling rate is not reached, grain size becomes

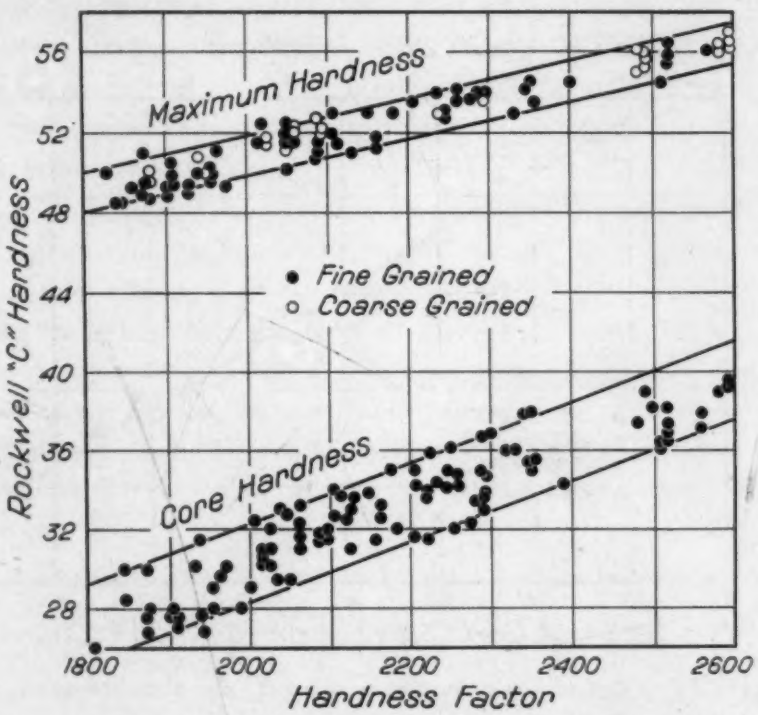


Fig. 6—Effect of Increasing Hardness Factor on the Maximum Hardness and Core Hardness of 1-Inch Round Bars Quenched from 1550 Degrees Fahr., and Tempered at 500 Degrees Fahr.

very important in determining surface hardness. The critical cooling rate for full hardness may not be reached due to

(1) intentionally slower cooling rate.
(2) the desired cooling rate is faster than the critical rate, but
  (a) gas films on the surface of the steel may retard the rate of heat transfer to such an extent that the critical cooling rate is not reached; and/or
  (b) If the scale does not break freely on quenching, the insulating effect of residual scale may act to retard the cooling rate.

No matter how the cooling rate is retarded the hardening reaction will behave in the same manner as indicated above for hardness penetration, i.e., will be a direct function of grain surface, and soft spots will occur where the critical cooling rate has not been reached. The hardness of the soft spots will depend on the analysis and grain size of the steel and on the effectiveness of the insulating spot of scale or gas.

The best method of avoiding soft spots is to be sure that the heating prior to quenching is carried out in such a manner that excessive scale formation is avoided, and to eliminate tenacious gas

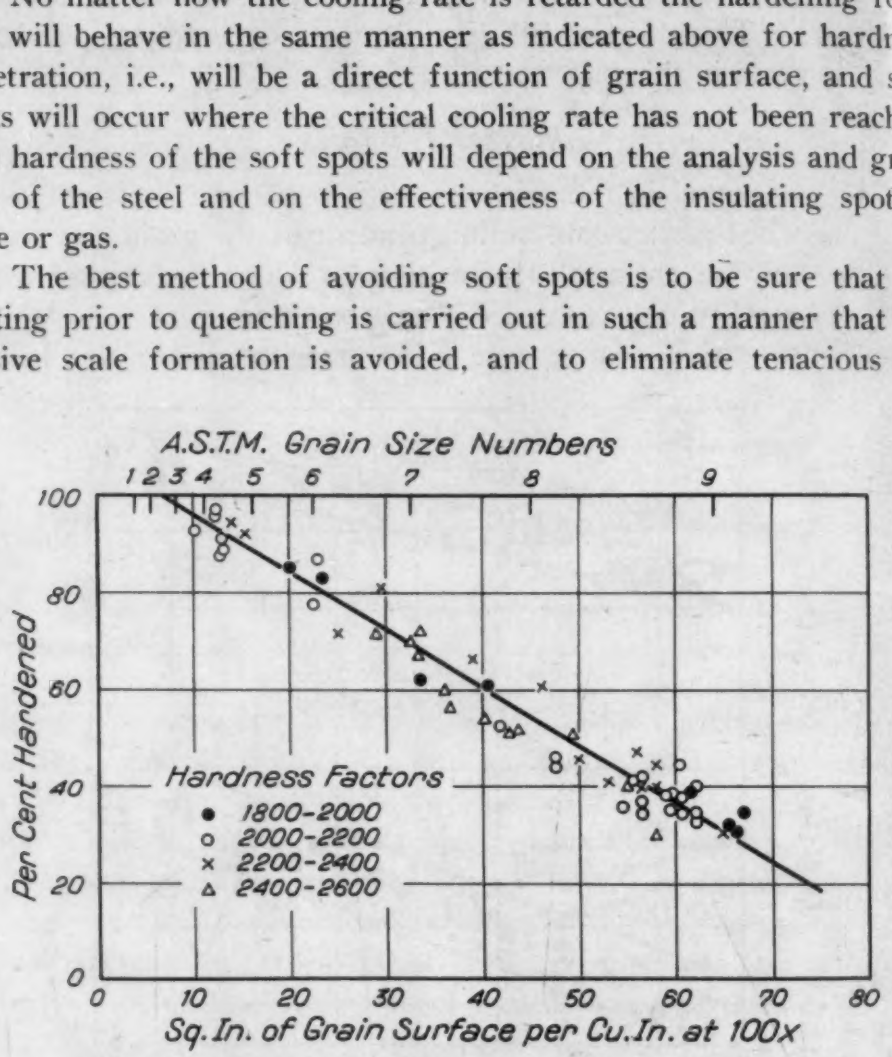


Fig. 7—Effect of Grain Surface Area on the Per Cent Hardened.

bubbles on the surface by proper agitation of the pieces or of the liquid quenching medium. In studying soft spots after quenching, a series of tests was made on fine-grained heats of S.A.E. 1045 steel in the form of 1-inch round bars, one set of bars heated in air, a duplicate set in a controlled neutral atmosphere. In the second or controlled atmosphere series, only a very light scale was formed on the pieces, whereas in the normally heated pieces the usual type of scale was present. Twenty Rockwell readings per inch of length were taken on each bar, and the percentage of soft spots as determined by these readings is shown in the tabulation appearing at the top of the next page.

Atmosphere	Per Cent of Rockwell Readings Showing Soft Spots	Total Number of Readings
Oxidizing	8.6	3210
Neutral	0.97	2585

The extreme importance of atmosphere control for reliable results in surface hardness can be seen from this table.

Experiments now in progress indicate that the type of steel may have a very definite effect upon the type of scale formed by a given atmosphere, thus affecting the manner in which the scale breaks away from the piece on quenching. This, therefore, affects the uniformity of surface hardness of the heat treated piece.

If the furnace atmosphere is oxidizing, or if the piece is improperly quenched, a coarse-grained steel, Type A, Fig. 2, is apt, because of a lower critical cooling rate, to give more uniform surface hardness than mixed and fine-grained steels, Type B, C, and D, Fig. 2, unless the latter are made in such a way that the scale has a tendency to break away cleanly from the metal surface during the quenching operation. If scaling can be closely controlled, the desired surface hardness can be obtained on steel of any grain size, provided that the critical cooling rate is exceeded in quenching. Under these conditions a fine-grained steel, a mixed-grain steel, or a coarse-grained steel will all give equally good results on surface hardness.

Summary

To summarize the effect of grain size on hardenability, it can be stated that hardness penetration is a direct function of the grain surface of the steel when this grain surface is designated as that present in the steel at the time of quenching. Steels of types B and C, Fig. 2, will give the same hardness penetration as steel D, Fig. 2, if quenched from 1550 degrees Fahr. (845 degrees Cent.) or lower. At higher quenching temperatures, steel D will be more shallow hardening than steels B or C, the penetration hardness being, as indicated in Fig. 7, a direct function of the grain surface of the steel. Changing the chemical composition of the steel within the range shown in Table VI, simply raises or lowers the actual hardness values at any given point on the diameter of the piece, without changing the relative penetration of hardness.

When the critical cooling rate is reached or exceeded, surface hardness is independent of grain size (or grain surface), but if for any reason the critical cooling rate is not reached, it is a direct func-

tion of grain surface, just as is the case for hardness penetration. The most common cause of soft spots is insulation of the surface by gas or scale films, which decrease the cooling rate to the point at which grain surface becomes a controlling factor in hardenability. Again in this instance the grain surface at the temperature of treatment is the important factor in determining the actual hardness and depth of the soft spots.

If steel is judged solely by a test carried out at temperatures and times different from those used in the heat treating operation, the interpretation of the observed grain size and, therefore, grain surface, may well be extremely misleading and the hardenability of the steel will not correspond to the observed grain size. Any of the methods described by Bain and Vilella will satisfactorily determine the austenitic grain size as it exists at any temperature of treatment.

Finally, it is evident from the foregoing that grain surface at the temperature of treatment is the essential factor in determining type of response to heat treatment. As the McQuaid-Ehn test gives no information as to actual grain size (grain surface) at any temperature other than 1700 degrees Fahr. (925 degrees Cent.) it should be obvious that its general use is not warranted in predicting the behavior of steels in heat treating operations.

Acknowledgment

The test data presented were obtained at the Cambria plant of the Bethlehem Steel Company. The assistance rendered by the members of the Metallurgical Department of that plant is gratefully acknowledged.

INFLUENCE OF ALUMINUM ON THE NORMALITY OF STEEL

By G. R. Brophy and E. R. Parker

Abstract

Pure iron-carbon-aluminum alloys were made, containing 0.001, 0.01, 0.10, and 1.0 per cent of aluminum. They were carburized in commercial carburizer and in an oxygen-free hydrocarbon atmosphere.

The cases obtained in hydrocarbons were all normal regardless of the aluminum content, and those in commercial compound, all abnormal. The depth of the abnormal layers decreased as the aluminum content increased.

The suggestion made by McQuaid, that small amounts of aluminum, as such, in solution in steel are responsible for the production of abnormal structures, is, therefore, untenable.

DURING the Campbell Memorial Lecture for 1935, H. W. McQuaid[1] suggested that small amounts of aluminum in solution in steel are responsible for the coalescence of carbides precipitated during the austenite transformation on cooling, and are therefore responsible for abnormal structure and fine grain.

This suggestion is made in spite of the fact that Bain[2] has shown for iron-aluminum alloys that, in the presence of 0.4 per cent of aluminum or more, abnormal structures are obtained only when carburization occurs in the presence of oxygen, and that perfectly normal structures are obtained when these same alloys are carburized in oxygen-free hydrocarbon gases.

Bain explains the action in the former case to be the result of the formation of aluminum oxide and its action on grain size.

If both views were correct it could be concluded only that the influence of aluminum is different, when present in amounts of less than approximately 0.1 per cent, than when present in larger amounts as reported by Bain.

[1]H. W. McQuaid, "The Importance of Aluminum Additions in Modern Commercial Steels", Transactions, American Society for Metals, 1935, Vol. 23, p. 797.
[2]E. C. Bain, "Factors Affecting the Inherent Hardenability of Steel", Transactions, American Society for Steel Treating, 1932, Vol. 20, p. 385.

A paper presented before the Eighteenth Annual Convention of the Society held in Cleveland, October 19 to 23, 1936. The authors are associated with the Research Laboratory, General Electric Co., Schenectady, N. Y. Manuscript received June 24, 1936.

This illogical conclusion led to a few critical experiments which are here reported.

A series of low carbon alloys were made, essentially free from oxygen, containing 0.001, 0.01, 0.10 and 1 per cent of aluminum, and were carburized in solid carburizing compound and in oxygen-free hydrocarbon gases.

Preparation of Samples

Electrolytic iron was first melted, held molten for approximately fifteen minutes, and then cooled in hydrogen. The ingots were then ground to remove the entire surface and charged, with 0.10 per cent carbon, into a crucible and melted in a high frequency vacuum furnace, where again they were held molten until gas-free. Suspended above the melt, and out of the hot zone, was the aluminum, or aluminum-iron hardener, as the case might be. After melting the iron the high frequency coil was raised so as to melt the aluminum which dropped into the molten iron. The coil was lowered again and the melt kept molten for about ten minutes to allow complete diffusion of the aluminum.

Carburizing

Carburizing was accomplished at 930 degrees Cent. (1705 degrees Fahr.) for six hours in the case of commercial solid compound, and for ten hours when hydrocarbon gas was used. The increased time in hydrocarbon was necessary in order to obtain sufficient depth of case. Both groups were cooled in the furnace at approximately the same rate.

The hydrocarbon atmosphere was obtained by bubbling especially purified hydrogen through benzene and then over the samples. Its freedom from oxygen is indicated by the fact that the hydrogen is capable of producing a bright anneal on stainless steel.

Results

Reference to the photomicrographs shows that those alloys carburized in oxygen-free hydrocarbon gas all have strictly normal pearlitic cases, but those carburized in compound show a high degree of abnormality in the hypereutectoid zone. Furthermore, this

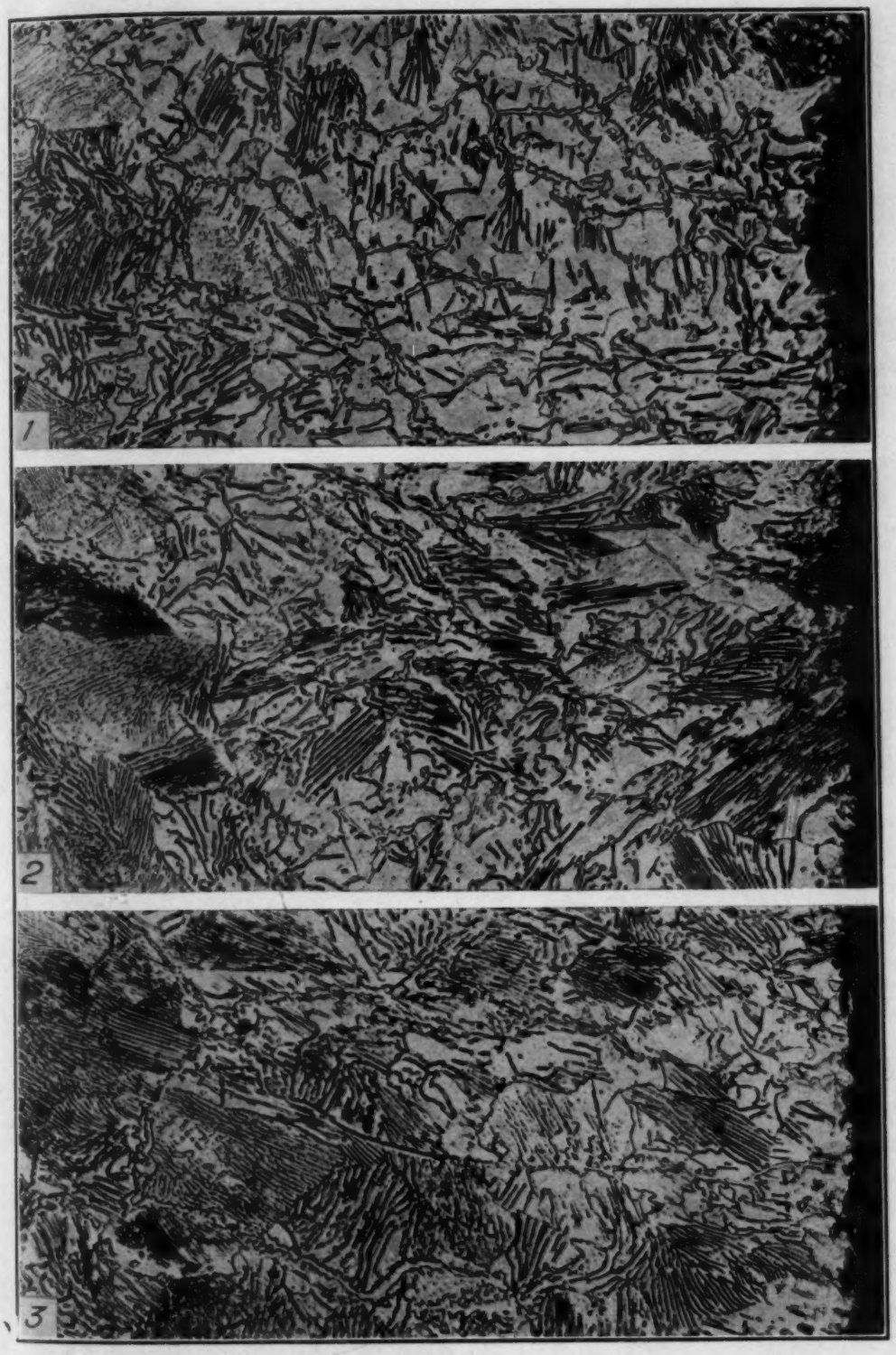


Fig. 1—0.001 Per Cent Al-Fe Alloy—Carburized at 930 Degrees Cent. (1705 Degrees Fahr.) for 6 Hours in Commercial Solid Compound. × 250.

Fig. 2—0.01 Per Cent Al-Fe Alloy—Carburized at 930 Degrees Cent. (1705 Degrees Fahr.) for 6 Hours in Commercial Solid Compound. × 250.

Fig. 3—0.10 Per Cent Al-Fe Alloy—Carburized at 930 Degrees Cent. (1705 Degrees Fahr.) for 6 Hours in Commercial Solid Compound. × 250.

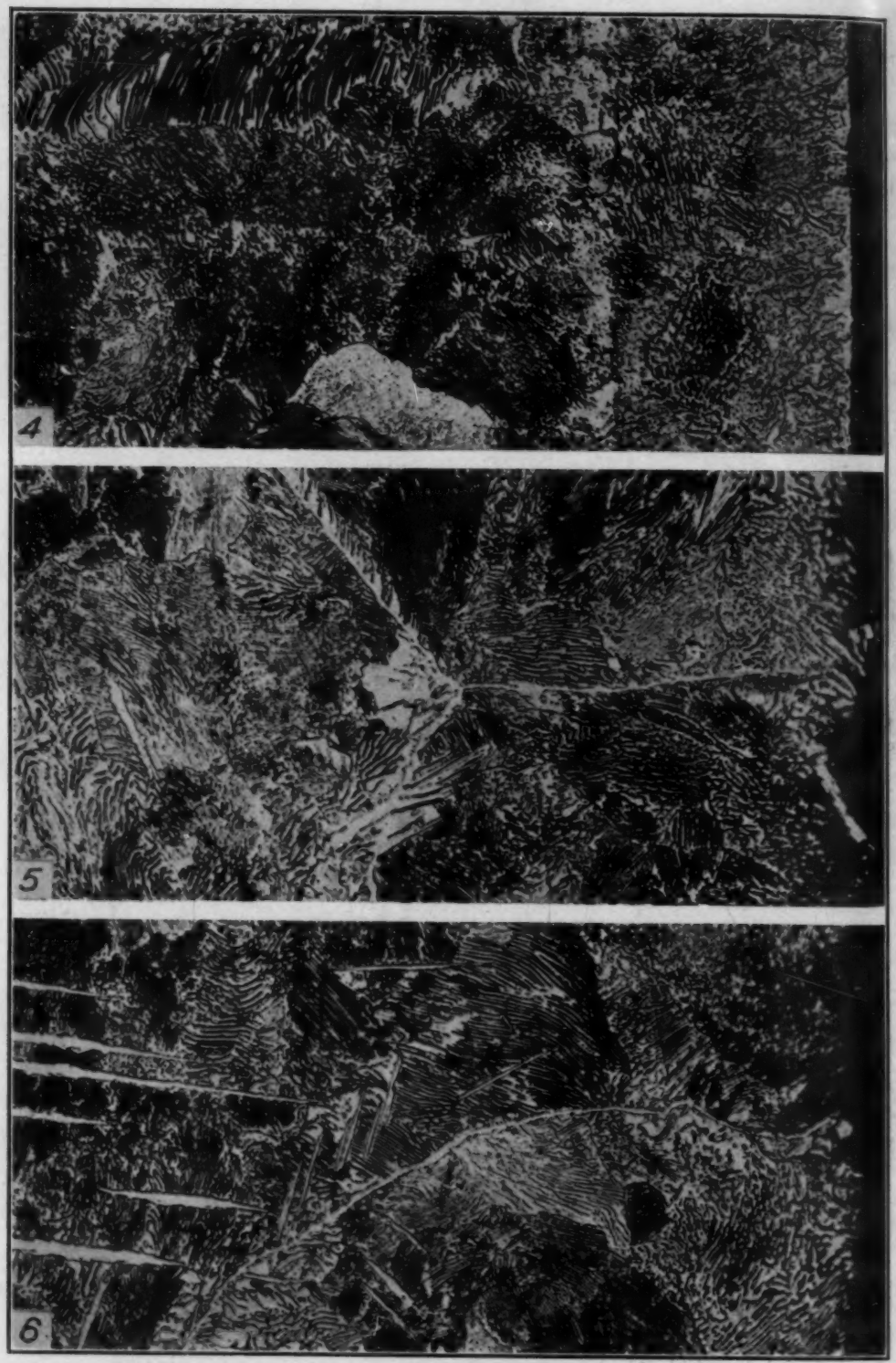


Fig. 4—1.0 Per Cent Al-Fe Alloy—Carburized at 930 Degrees Cent. (1705 Degrees Fahr.) for 6 Hours in Commercial Solid Compound. × 250.

Fig. 5—0.001 Per Cent Al-Fe Alloy—Carburized at 930 Degrees Cent. (1705 Degrees Fahr.) at 10 Hours in Hydrocarbon Gas. × 250.

Fig. 6—0.010 Per Cent Al-Fe Alloy—Carburized at 930 Degrees Cent. (1705 Degrees Fahr.) for 10 hours in Hydrocarbon Gas. × 250.

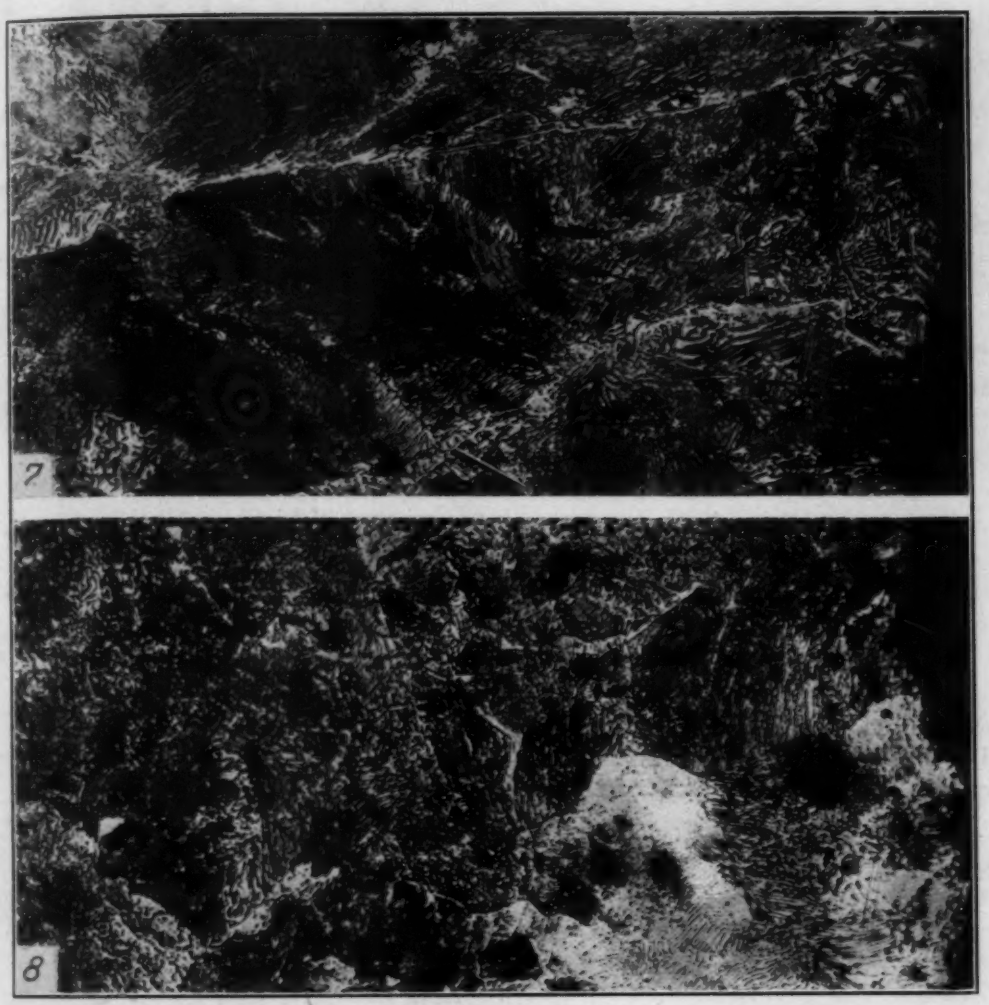


Fig. 7—0.10 Per Cent Al-Fe Alloy—Carburized at 930 Degrees Cent. (1705 Degrees Fahr.) for 10 Hours in Hydrocarbon Gas. × 250.
Fig. 8—1.0 Per Cent Al-Fe Alloy—Carburized at 930 Degrees Cent. (1705 Degrees Fahr.) for 10 Hours in Hydrocarbon Gas. × 250.

zone was characterized by numerous fine particles visible on the unetched surface at high magnification. These particles were entirely absent in the samples carburized in the hydrocarbon atmosphere.

Photomicrographs 1 to 4 show the case structures of the samples carburized in commercial solid compound. At first glance it might be assumed that the sample containing 0.001 per cent of aluminum is abnormal to the greatest degree and that the high aluminum alloy is the least abnormal. This, however, is only apparent.

It has been shown by Grossmann[3] that oxygen is absorbed during carburizing in solid compound, and by Bain[1] that the aluminum

[3]M. A. Grossmann, "Oxygen in Steel," TRANSACTIONS, American Society for Steel Treating, 1930, Vol. 18, p. 601.

present in steel probably combines with the oxygen to form the visible particles previously mentioned. This being the case, in the presence of small amounts of aluminum, oxygen will penetrate to considerable depths combining with all of the aluminum it meets. If, on the other hand, aluminum is present in high concentration, as in the 1 per cent aluminum alloy, the available oxygen will be fixed so rapidly that the depth of penetration will be small.

It is in these zones only, where aluminum and oxygen have combined, that abnormal structures are found. These zones do not differ materially except in depth as follows:

0.001 per cent Aluminum = 0.018 inch
0.01 per cent Aluminum = 0.015 inch
0.10 per cent Aluminum = 0.012 inch
1.00 per cent Aluminum = 0.0045 inch

Conclusions

McQuaid's suggestion that traces of aluminum in solution, as such, cause abnormality is not confirmed, but to the contrary, Bain's hypothesis is supported.

Aluminum in extremely small amounts is sufficient to cause abnormality if oxygen is available.

The depth of penetration of oxygen and, therefore, the depth of the abnormal layer, decreases with increasing aluminum content.

Carbon penetrates beyond the high oxygen, abnormal layer to form normal carbides and pearlite.

DISCUSSION

Written Discussion: By O. W. McMullan, Youngstown Sheet and Tube Co., Indiana Harbor, Indiana.

It would be an interesting addition to the paper to have had the results shown on a steel free from aluminum. It is the writer's opinion that the same difference in normality in such a steel would have been shown between samples carburized in commercial carburizers and oxygen-free hydrocarbon gases. The fact that 0.001 per cent aluminum apparently shows greater rather than less normality points in this direction, as does also the fact that several times that much aluminum may have no visible effect on commercial steels. Bain[1] has actually shown that very pure iron may be abnormal when pack carburized but normal when carburized in pure hydrogen and hydrocarbon. The authors have therefore shown that it is oxygen rather than aluminum that promotes abnormality. In fact, the effect of aluminum is exactly the contrary; it promotes normal-

ity in steel and any apparent effect shown toward abnormality is secondary rather than primary.

Normality or abnormality in steel, if by abnormality is meant coalescence of carbide and separation of free ferrite, is a function of composition and cooling rate. Any of the common steels may have their apparent normality or abnormality greatly changed by varying the cooling rate after carburizing. Steels containing alloying elements may be cooled much slower than plain carbon steels without showing abnormality because of slower critical cooling rates and suppressed transformation points. Thus a plain carbon and a nickel-chromium steel might be cooled at such a rate as to develop a normal pearlitic structure in the plain carbon steel and a martensite or even austenitic one in the alloy steel. Should the steels be cooled slow enough to develop normal pearlite in the nickel-chromium steel, coalescence of carbide and abnormality may appear in the carbon steel. The usual procedure in making McQuaid-Ehn tests does not take into consideration this difference in composition but it is not necessary to do so as far as practical results are concerned. The addition of small amounts of aluminum to commercial steels would not show the abnormal structures present in Figs. 1 and 2, but rather thickened and broken carbide boundaries. This difference is perhaps due to the lack of manganese or other hardening elements in the author's steels. The difference in structure between Fig. 1 and Fig. 4 is quite probably not caused by oxygen alone. Aluminum as a metal is a strong hardening element and the higher aluminum suppresses the critical cooling rate and promotes normality. The abnormality of Fig. 4 occurs not because aluminum is present but only near the surface where excess oxygen has been absorbed. Note also the difference in pearlite structures between Figs. 5 and 7 or 8.

Since degree of normality is a result of carbon diffusion it may be affected by a substance which changes the rate of diffusion or changes the distance of travel by changing the grain size. Aluminum works in opposite directions by combining with oxygen which appears to reduce abnormality but at the same time forms Al_2O_3 to which has been assigned the cause of fine grain size and tendency toward abnormality.

Primarily, aluminum promotes normality and a coarse grain, but when both aluminum and oxygen are present in the right amounts a fine-grained structure results, which in turn, in steels of low alloy content and high critical cooling rates, may result in abnormality in the McQuaid-Ehn test. This puts considerable restriction on the authors' statement that aluminum in extremely small amounts is sufficient to cause abnormality if oxygen is available.

Written Discussion: By Stephen F. Urban, Carnegie-Illinois Steel Corporation, Chicago.

The paper by Messrs. Brophy and Parker again confirms the work of Grossmann and Bain on the effect of oxygen absorption, during pack carburizing, on so-called abnormality in steels containing aluminum. The authors have further demonstrated that steels containing aluminum do not exhibit abnormality when carburized in a hydrocarbon atmosphere free of oxygen.

Since most of the published work is confined to the effect of aluminum on

[1]Reference (2) of paper discussed.

abnormality, during pack carburizing, it is felt that we should bear in mind the case abnormality in ingot and electrolytic iron. That is, the oxygen penetration theory, during pack carburizing (at 1700 to 1725 degrees Fahr.), may not explain abnormality in ingot or electrolytic iron, particularly since the authors assume that abnormality in their steels was caused by the reaction of absorbed oxygen with aluminum. It should also be observed that steels deoxidized with silicon may be entirely normal after pack carburizing, even though it is assumed that oxygen absorption occurs in this process. It therefore appears that oxygen absorption may not be a complete answer to the problem of abnormality.

The above discussion is not intended to detract from the contribution made by the authors, but it is made with the thought that it may stimulate more work on the mechanism of abnormality.

Written Discussion: By H. W. McQuaid, Republic Steel Corporation, Massillon, Ohio.

It is difficult for me in studying this paper to understand by what method of reasoning the authors arrived at the conclusions they did from the evidence submitted. It would have been much more conclusive if a definitely hypereutectoid case had been obtained in both methods of carburizing. It is possible that if a definite hypereutectoid case had been developed by carburizing in a benzene mixture that the structure would show considerable evidence of divorcement. That this is the case may not necessarily be true, due to the presence of the hydrogen which might in itself have some effect upon the structure obtained upon cooling. It is a well known fact that when the solution of the carbon in the austenite takes place at such a rate that practically eutectoid composition is developed, that the structure is, as a rule "normal." Thus it would usually be found that the zone of eutectoid in a carburized steel rarely shows any evidence of divorcement.

As far as I can see, this paper checks the work of Houdremont who showed that iron as pure as possible in its composition was abnormal in structure. Certainly no one would classify steel shown in Fig. 1 as a steel containing enough aluminum to be classed as an aluminum-treated steel, and yet, the indications are that this steel shows more divorcement than any of the others.

While the work done by Dr. Grossmann in 1930 would indicate that oxygen was absorbed during carburizing by means of a compound, it has been found that this is not necessarily always true and most of the attempts which we have made to check this result have not indicated that oxygen is absorbed during carburizing in a compound. In fact, results seem to indicate that the contrary is usually true. It is difficult to conceive of oxygen being absorbed in steel treated in a compound because the act of carburizing in itself requires the absorption of carbon which is in itself a reducing reaction. It is a well known fact that compound carburizing depends on the presence of charcoal, which has a double function to perform. First, it must convert the carbon dioxide available from the energizer into carbon monoxide and second, it must remove from the surface of the work any carbon dioxide (or oxygen) which is present. This is entirely an adsorptive phenomenon and unless the charcoal is preferentially adsorptive to the carbon dioxide the carburizing is unsuccessful. It is for this reason that we cannot carburize using carbon in the form of coke, soot or graphite, and it is also the reason why pure carbon monoxide is not a satis-

factory carburizing medium. It is only the fact that certain types of charcoal will adsorb carbon dioxide in preference to carbon monoxide at the carburizing temperatures that makes compound carburizing possible.

The results shown in this paper are very interesting and indicate the need for further study as to the causes of the pearlite divorce in the purer steels and also indicates that if the absorption of oxygen takes place in carburizing with compound, that we have yet to explain why abnormal steels are not produced in every case when carburizing in compound. We know that without the presence of any aluminum it is possible to produce the abnormal structure by carburizing an over-oxidized steel.

It is possible that the cause of " abnormality" in the purest steels is due to the very high rate of austenite transformation and the relatively high temperature at which it occurs. This is, it seems to me, a fairly logical explanation of the cause of "abnormality" in the steel shown in Fig. 1.

Written Discussion: By C. L. Shapiro, Halcomb Steel Co., Syracuse, N. Y.

The authors should be congratulated upon their excellent and pertinent paper which tends to explain whether or not aluminum in solution is directly responsible for abnormality in carburized steels. And, it is unfortunate that they made their paper so brief, as they might have cleared up a few questions that their photomicrographs raise; namely:

(a) Since the carburizing periods in the solid compound and hydrocarbon gas are 6 and 10 hours respectively, and since the austenitic grain size of the latter is greater than that obtained by carburization in the solid medium, is this variation in grain size due to the difference in the carburizing time or to the carburizing atmosphere? If the difference in grain size is due to the carburizing atmosphere, to what may it be attributed?

(b) If the carburizing time was constant in both methods, will the amount of aluminum in the steel affect the austenitic grain size rating in the hydrocarbon gas as much as in the solid carburizing compound?

Authors' Closure

The inclusion of the results of carburizing on aluminum-free iron, as suggested by Messrs. McMullan, McQuaid, and Urban, would have greatly enhanced the value of our work without a doubt. This should have been done. However, as stated in the paper, we believe that there is no difference in the degree of abnormality in the pack carburized cases, but only in the depth to which the abnormality extends, which, we believe, is the extent of the oxygen penetration. Beneath the abnormal zones of the higher aluminum alloys, normal hypereutectoid structures were noted even though the micrographs presented do not illustrate this point.

Contrary to Mr. McMullan's interpretation, we have not shown that oxygen alone is responsible for abnormality but rather that in this case the combination of oxygen with aluminum to form a dispersed insoluble solid is the cause. Other insoluble solids, such as vanadium carbide, have been found to act similarly.

A convincing bit of evidence in support of the insoluble solid theory is the fact that the surfaces of our small ingots which were in contact with the

alumina crucibles and were contaminated thereby showed abnormal cases even when carburized in hydrocarbon gas. Our samples were quarter sections cut from the ingots, so that each had both contaminated and uncontaminated surfaces.

Mr. Shapiro's questions are partly answered by Mr. McMullan's discussion of the influence of aluminum on the grain size and normality. It may be said, however, that the additional time in the hydrocarbon gas has increased the grain size somewhat but we must not overlook the role played by the aluminum oxide formed during pack carburizing in refining the grain of the hypereutectoid zone. This, it is believed, is the reason for the difference in grain size found in the two methods.

Carburizing for six hours in the hydrocarbon gas does not give the same carbon content in the case—in fact less than eutectoid carbon is obtained, so that the grain sizes cannot be compared with those obtained by the pack method.